METALLURGIA

The British Journal of Metals

(INCORPORATING THE METALLURGICAL ENGINEER)

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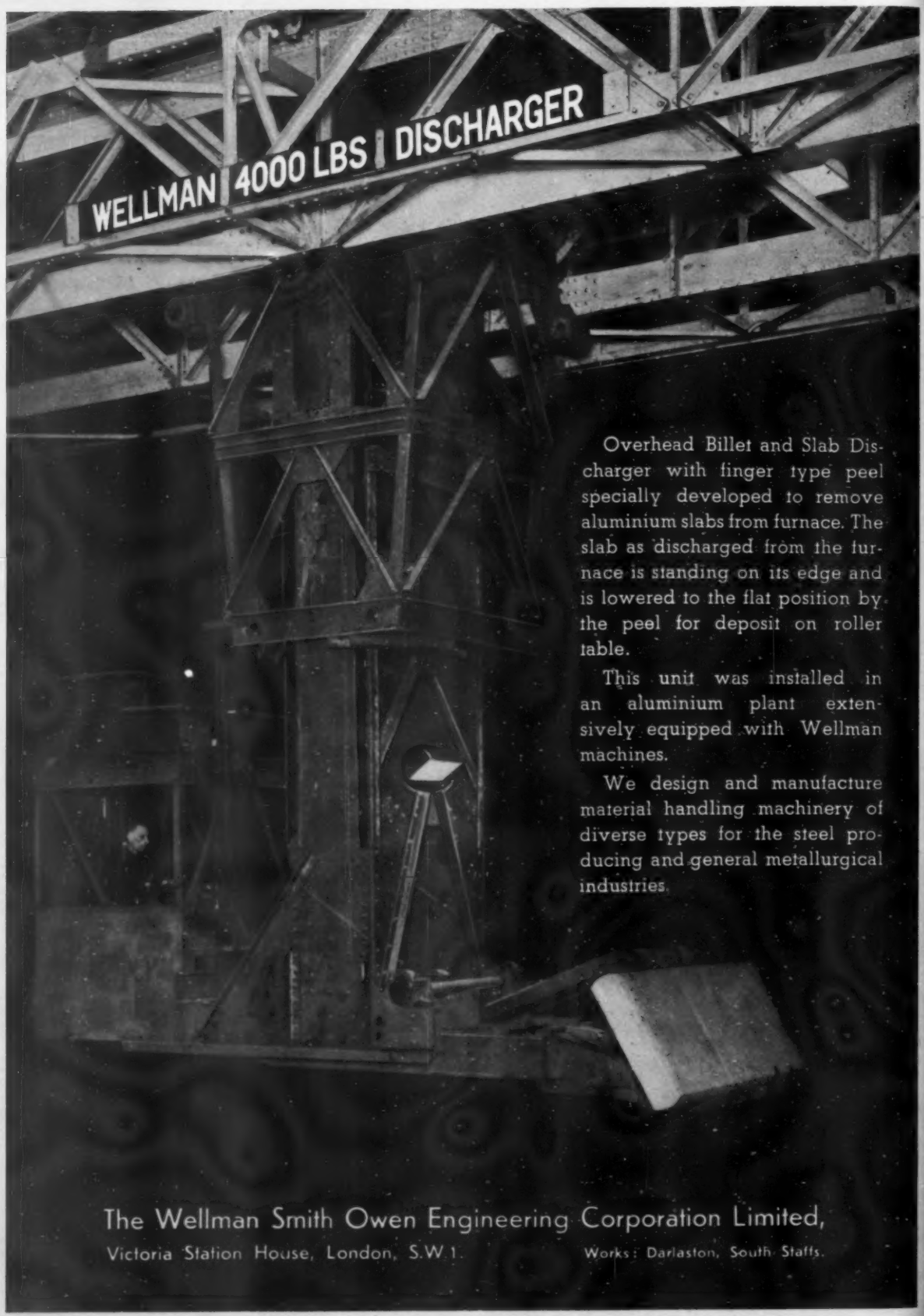

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METALLURGIA

THE BRITISH JOURNAL OF METALS.

INCORPORATING "THE METALLURGICAL ENGINEER."

JUNE, 1946 VOL. XXXIV. No. 200

What is Britain's Aluminium Policy?

THE recent reduction in the price of virgin aluminium in ingot or notched bar form, made possible as a result of a contract made by the Ministry of Supply for the purchase of 215,000 short tons from Canada over this and next year, has directed attention to the position of the aluminium industry, especially with regard to its future in Britain. The fact that the international aluminium cartel had ceased to function adds interest to the recent announcement of cheaper virgin metal, but, up to the time of writing, there is no indication of a definite Government policy regarding primary aluminium likely to guide manufacturers of aluminium goods and to assist them in determining possible directions for development in home and overseas markets that will not conflict with the nation's need for economy.

Reviewing the position of the industry, it will be generally appreciated that, during the past ten years, it has been subjected to many influences tending to upset its normal progress. A brief summary of some of the more important of these influences will indicate the position. First, there was the enormous artificial development of German production for the purpose of replacing imported copper wherever possible by indigenous aluminium and at the same time satisfying the rapidly increasing demands for aircraft construction in preparation for war. Then came the impact of rearmament on our own industry which resulted in an expansion of British light alloy fabricating capacity to a degree far in excess of immediate potential normal consumption. Both these factors were in themselves sufficient to upset the even trend of development under the control of the aluminium cartel; but since the outbreak of war the far greater increases in production capacity of both primary metal and fabricated products in the United States and Canada have provided the industry with fresh political problems which are further complicated by the successful prosecution of the Aluminum Company of America under the anti-trust laws, the proposed complete liquidation of the German light alloy industry and the need in Britain to restrict imports not absolutely essential.

During the period of forced expansion of aluminium consumption in Germany the metal was employed wherever possible and considerations of utility or cost were not permitted to prevail. In the United States, however, the Aluminum Company of America subjected every new application of their metal and its alloys to critical investigation with the object of ensuring as far as possible that their increasing employment would be developed only where technical or economic superiority could be established. The magnesium producers in this country have been careful to restrict sales promotion to applications where the properties of their material offered special advantages, but in the aluminium industry there have been instances where the replacement of other materials by aluminium was advocated, although considerations neither of cost nor technical suitability provided a reason for the substitution. It is probably true that these instances have been caused by over-enthusiasm of sales representatives, but they need to be guarded against in building up strong and healthy aluminium industry. The inherent advantages offered by aluminium are its lightness, strength, attractive appearance and resistance to some forms of corrosion. To offset these characteristics has been its relatively high cost, but when considering cost the low specific gravity of the metal must be taken into account.

The war has left the United States and Canada by far the largest and most economical primary aluminium producers in the world. The Aluminum Company of America is still the most efficient smelter in the United States and has a capacity far in excess of probable demands in the near future. The Aluminium Company of Canada has its ore treatment plant at Arvida, situated on the banks of Saguenay river, and deep water wharves at Port Alfred, about 20 miles away, on Ha Ha Bay, provide direct access by sea, excepting during the winter when the river is frozen over, to both foreign markets and the bauxite deposits in British Guiana. This Company's principal reduction plant for the production of aluminium ingots is also located at Arvida. The relatively low cost of production and transport would permit metal from Canada being landed in Great Britain at a price far below the actual cost of primary metal extraction in this country. In 1944, the production of primary ingot in Canada totalled nearly 500,000 short tons, only 6% of which was consumed in Canada. Great Britain was the largest buyer, taking approximately 55% of the production.

It is apparent that the productive capacity in North America is such as to warrant the greatest possible development of the application of aluminium to industrial and general use. In Britain the position is not so straightforward. The output of the British Aluminium Company, who are, for practical consideration the only domestic producers, is in the region of 30,000 tons per annum. A higher rate of output has been obtained, but having regard to the occasional failure of the comparatively small water supply to the hydro-electric generating station, the figure given is reasonably accurate. Any consumption in excess of domestic production must, therefore, be catered for by importing metal. It is important to British economy, therefore, that applications for aluminium should be strictly utilitarian until our finances are in a healthier state. At the present time the shortage of steel and cast iron is providing an artificial stimulus to the use of aluminium and probably

the contract with Canada is to meet this demand, although the stock of secondary aluminium would seem to be adequate for the purpose. Many of the uses to which aluminium is being now put, because of shortages of iron and steel, doubtless will become permanent and this would constitute a definite gain for the aluminium industry.

The future trend of British domestic consumption, however, is clearly dependent on supplies and cost. Concerning the former, it has been pointed out above that all metal in excess of approximately 30,000 tons a year must be imported and the most likely source of supply is Canada. The price of aluminium hitherto has been controlled by the cartel and its successor, Alliance Aluminium Cie, Basle. The latter has now been liquidated at the instance of Aluminium Limited, as a result, doubtless, of the findings of the United States Courts in the action against the Aluminum Company of America under the anti-trust laws, also the fact that the liquidation of the German smelters leaves the American producers in a position of such overwhelming superiority as to make them virtually heedless of any European competition. Already American prices of many semi-finished products are considerably lower than in this country and further reductions may be expected.

At present the Government is the only buyer of primary aluminium ingot in this country; but, if the market were opened to free competition from North America, the British Aluminium Company would probably be unable to compete and would be compelled to rely for revenue on the products of their rolling mills and other fabricated products. Even here they would be faced with competition from the American-owned Northern Aluminium Company. In this connection it is noteworthy that the relatively highly efficient rolling mill and extrusion works at Rogerstone in Monmouthshire are owned by the Government and are at present leased to the Northern Aluminium Company. Prices of most semi-finished products were formerly controlled by the Wrought Light Alloys Association, but, during the hearing of the action in the United States Courts mentioned above, the Northern Aluminium Company was instructed by the parent company to withdraw from price agreements with the Association. Excepting at Rogerstone and in the new mill erected by the British Aluminium Company, there is no hot rolling equipment in Britain capable of breaking down the large cast slabs now in general use in the United States. All the British mills are designed for jobbing and none is equipped for continuous strip rolling. It is true that equipment schemes have been considered, but when they will be proceeded with, and to what extent, will depend largely on the policy of the Government.

It would seem from the foregoing that the future progress of the British aluminium industry is governed by the desirability or otherwise of importing metal from Canada. The Government has an interest in the reduction plant at Arvida, since the extensions made during the war were paid for from British funds, and it is probable that the contract with Canada for this and next year also embraces some form of agreement for future supplies. But the position is not by any means clear. The supply available from the British producers is about 30,000 tons of relatively high priced metal. In the national interest it is desirable to maintain this supply, but, while every encouragement should be given to the development of the industry so that it can compete with suitable aluminium products in the world's markets, there is a need for clarifying the position. Is the lead already given by the Government, in arranging the contract with Canada, an indication of future policy to reduce the price of primary metal so that semi-finished products will be available at a competitive level? Can the industry proceed to develop on this assumption?

The British Non-Ferrous Metals Federation

AT the first annual general meeting of the above Federation, held on June 6, in Birmingham, the President, Mr. Horace W. Clarke, in presenting a Report on the progress of the Federation since its inception, payed tribute to the excellent work of its officers and committees. He welcomed the revival of the Non-Ferrous Wrought Metals Export Group, under the Chairmanship of Mr. R. Finch, which embraced all firms in the Industry whether members of the Federation or not.

One of the most important developments of the first year's work of the Federation had been the establishment of a Statistics Department, which had taken over from the Ministry of Supply responsibility for certain non-ferrous metals statistics. It was hoped that this work would shortly be extended in various directions. There could be no doubt that in modern conditions accurate and speedy statistical information was essential to the planning of Industry. Before the war, non-ferrous metals statistics in this country had lagged a long way behind those of Germany and the United States of America. It was hoped that the Federation Statistics Department would increasingly fill this want and that with the progress of time its activities would embrace a wider field to the lasting benefit of all concerned.

In considering its post-war plans, the Federation had become very conscious of the vital importance to the Industry of the twin pillars of Technical Education and Scientific Research. The Non-Ferrous Metals Industry had a very proud record in both these respects, which the Federation hoped to continue. Mr. Clarke reminded the meeting of the recent contribution from Trade Associations, individual Firms and other representatives of the Industry amounting to £138,000 for the establishment of a Chair of Industrial Metallurgy at Birmingham University, now held by Dr. Leslie Aitcheson. On the side of Scientific Research, Trade Associations and individual Firms had, for the last 25 years, been staunch supporters of the British Non-Ferrous Metals Research Association, and arrangements have now been made for the Federation to make an inclusive annual subscription to the Research Association on behalf of all its members. Mr. Clarke said that after a lifetime spent in the Non-Ferrous Metals Industry he was more than ever convinced that the two prime essentials for the well-being of the Industry and the maximum benefit to the country were high-grade technical education, in order to provide a constant flow into industry of highly qualified technicians, and also an intensive scientific research both long- and short-range. The declared policy of the Federation was to extend both research and the training of likely young men in the higher branches of metallurgy.

The Supersonic Method for the Detection of Internal Flaws

An Examination of the Method as Applied to Aluminium Alloy Billets and Semi-Finished Components

By E. G. Stanford, B.Sc., A.Inst.P., and H. W. Taylor, B.Sc.
Research Division, Aluminium Laboratories Limited, Banbury.

The supersonic flaw detector has been used for the routine inspection of aluminium alloy extruded materials and for a detailed examination, in the laboratory, of specimens containing internal cracks. Remarkably good results have been obtained on rectangular sections and cylindrical bar, but extruded sections of complicated shape are difficult to deal with. A scheme is outlined whereby the apparatus may be assembled for routine testing under production conditions.

Introduction

THE work described in this article was carried out on behalf of the Northern Aluminium Company, Ltd., in 1944, and subsequently, and covers an investigation of the field of application, to aluminium alloy billets and sections, of a supersonic apparatus designed and supplied by Messrs. Henry Hughes & Son, Ltd., for the detection of internal flaws.

The Principle of the Supersonic Flaw Detector

In order to promote a comprehensive understanding of the text of this article, a brief outline is given of the principle of supersonic flaw detection and the apparatus used. For a more complete account the reader is referred to the literature on the subject.[1, 2, 3, 4, 5, 6, 7]

The application of sound waves to the detection, and accurate location, of internal flaws in materials has been made possible by the use of high frequency waves known as "supersonic" waves. The frequency of these waves (2½ mega-cycles was the frequency used for the work described here) is well above the audible range, they are capable of being transmitted by most materials and their wavelength is short enough to secure rectilinear propagation.

In the case of the flaw detector, the supersonic waves are produced by the utilisation of the piezo-electric effect of quartz. When an alternating electric potential is applied to a crystal of quartz an expansion of the crystal takes place in one-half cycle and a contraction in the other. Conversely, when a crystal of quartz is subjected to an alternating mechanical pressure an alternating electrical potential is generated across it. In this manner the energy of an electrical impulse is converted into a damped train of supersonic waves which are injected into the material under inspection by holding the quartz crystal (the transmitter) against the surface of the material. The waves travel through the material to the opposite boundary, where they are reflected back to another crystal (the receiver) placed against the surface alongside the transmitter. The receiver converts the energy of the supersonic waves into electrical energy which is subsequently amplified and made visual by means of a cathode-ray tube. If the material contains a crack situated in the path of the supersonic beam, then reflection of some of the supersonic energy takes place at the crack and this reflected energy (an echo) reaches the receiver before the echo from the boundary (referred to as "the bottom echo") of the material. The difference in times of arrival, at the receiver, of the echo from the crack and the bottom echo is indicated on the time sweep of the cathode-ray tube, thus providing a means of observing the presence and location of the crack.

Fig. 1.—General view of supersonic apparatus.

The Supersonic Flaw Detector

The general appearance of the equipment is shown in Fig. 1, and Fig. 2 is a schematic diagram of the parts of the equipment. The trigger unit first supplies a

1 L. Bergmann, "Ultrasonics," London, 1938, G. Bell & Sons, Ltd.
2 C. H. Desch, D. O. Sproule, W. J. Dawson, "The Detection of Cracks in Steel by Means of Supersonic Waves," Iron and Steel Institute, Annual General Meeting, 1st May, 1946.
3 F. A. Firestone, *Metal Progress*, 1945, September, 505.
4 F. N. Simons, *Metal Progress*, 1945, September, 513.
5 D. O. Sproule, *Machinist*, 1945, 89, 2013–2016.
6 J. V. Russell, H. E. Pellitt, *Iron Age*, 1946, 157 (5), 38.
7 W. S. Erwin, *Steel*, 1945, vol. 116, No. 10, p. 131.

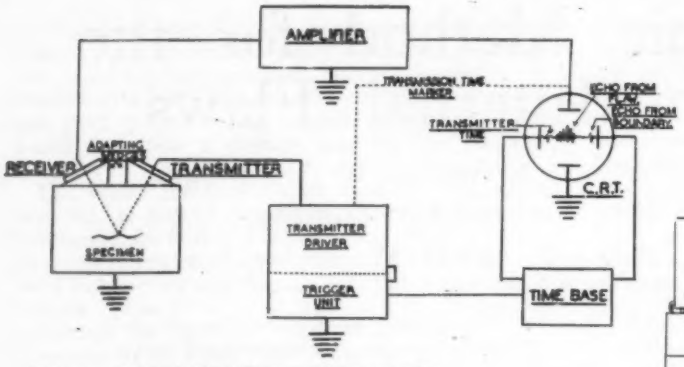

Fig. 2.—Schematic diagram of supersonic apparatus.

triggering electrical impulse to the time-base and starts the spot on its way across the screen of the cathode-ray tube ; a short time later it supplies a similar electrical impulse to the crystal-transmitter. The crystal-transmitter consists of a quartz crystal in contact with an aluminium wedge, as shown in the diagram ; the wedge directs a beam of supersonic energy into the region of the specimen which is to be tested for flaws. The receiver, which is a similar crystal to that of the transmitter, is mounted, in a like manner, on another aluminium wedge and picks up the energy reflected from the bottom of the specimen or from any crack which may be situated in the material of the specimen. The receiver-amplifier is arranged to give a vertical deflection to the cathode-ray spot at the moment of arrival of the reflected beam, or echo. The time-base gives an approximately uniform velocity to the spot, from left to right, and the speed of its traverse is adjusted so that it reaches the right hand side of the screen at the moment the first bottom echo reaches the receiver (see Fig. 5). If a crack is present in the specimen, an echo from this crack will reach the receiver before the bottom echo and an indication will appear on the tube trace as shown in Figs. 6(b), 7(b) and 9(b).

Introduction to the Present Investigation

The approach to this investigation was made with two main objects in view, namely :—(1) the development of a scheme whereby the apparatus could be used in production on a routine inspection basis, and (2) the examination, in the laboratory, of the results obtained from the apparatus when used to explore aluminium and aluminium alloy billets and various types of semi-finished components.

The present equipment has achieved remarkable success on rectangular sections and cylindrical bar of 6 ins. maximum diameter. More recent models of the same apparatus which have been designed on the basis of experience gained by the use of this equipment are capable of working ranges from $\frac{1}{2}$ in. to 12 ft. or from 6 in. to 25 ft. This development is of some importance to the aluminium industry since it frequently allows sections to be examined from one end and permits inspection of the largest size of cast billets (18 ins. diameter).

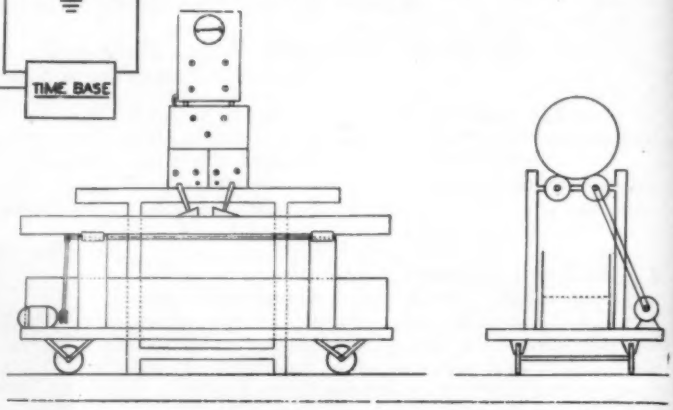

Fig. 4.—Assembly for routine inspection of extruded cylindrical bar of 6 in. maximum diameter.

Throughout this investigation the procedure adopted was that of assembling the apparatus near to a production unit and inspecting a certain percentage of the material produced. Whenever the presence of a defect was indicated, the portion of material containing that defect was taken to the laboratory for more detailed examination. Here the form taken by the trace on the cathode-ray tube was thoroughly examined and the specimens were then marked and cut up at positions which, it was estimated, would reveal the defect. Slices of the sections, rectangular and circular, were faced and caustic pickled to show up any defects, the type and position of which were compared with the estimates made from the indications given by the apparatus.

The photographs included in this article, and to which reference is made later, are of some of the defects found in this manner, and the diagrams accompanying each of the photographs are drawings of the actual traces on the cathode-ray tube which indicated the presence of the defects concerned.

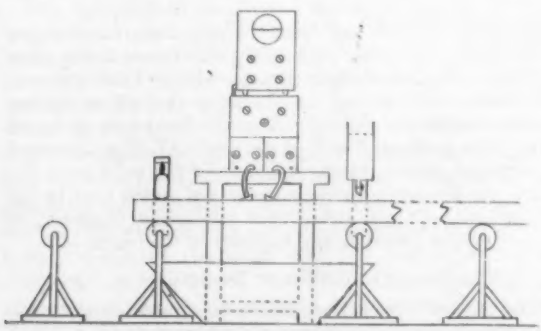

Fig. 3.—Assembly for routine inspection of rectangular extruded sections.

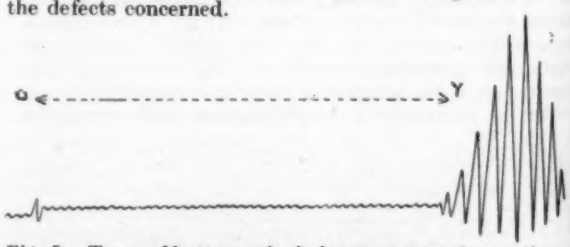

Fig. 5.—Trace of bottom echo in large rectangular section. No interference from cracks.

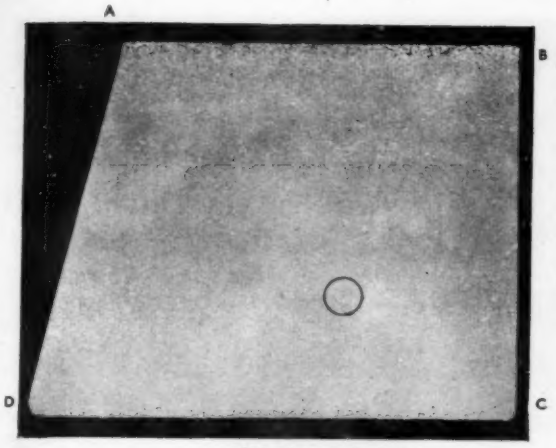

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Fig. 6a.—Small crack indicated by circle in B.S.S. A 2L.40 spar. Note three sub-surface defects along bottom edge, of the same order of size as the crack detected. ×1.

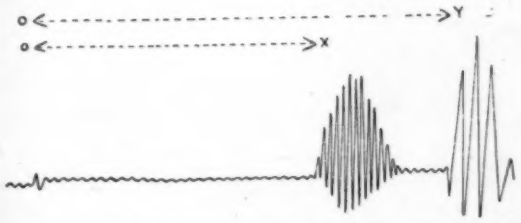

Fig. 6b.—Trace indicating presence of small crack shown in Fig. 6a.

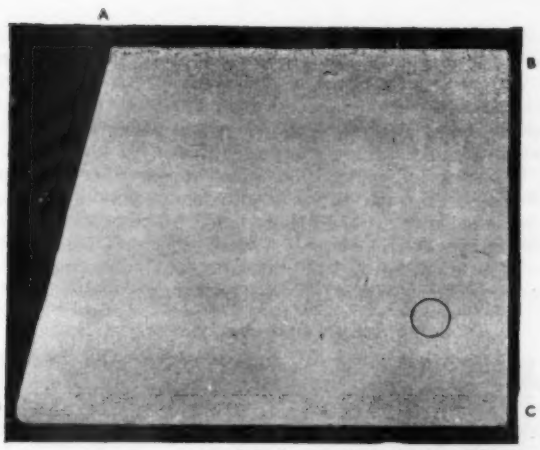

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Fig. 7a.—Small crack indicated by circle in B.S.S. A 2L.40 spar. ×1.

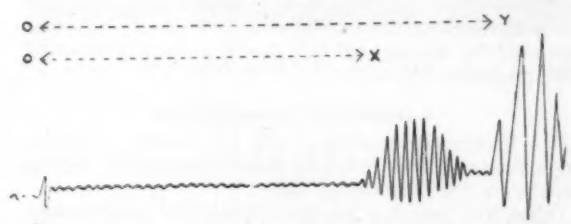

Fig. 7b.—Trace indicating presence of small crack shown in Fig. 7a.

Preparation of the Surface of the Material Undergoing Inspection

The condition of the surface of the material has an important bearing on the degree of success achieved with the present apparatus. Almost perfect contact must be made between the transmitter-receiver unit and the surface of the material. At the moment this can be achieved in the case of very rough surfaces by machining and covering the machined surface with a suitable medium, such as ordinary machine oil, which transmits the supersonic waves.

It has been found that an extruded surface requires no machining operation prior to inspection. With this apparatus it is possible, though not desirable, to work on a sawn surface but a rough cast surface must be machined; this may be done with a hand grinding wheel.

Prior to inspection, all types of surface must be covered with a film of medium capable of transmitting the supersonic energy from the crystal wedges to the material. Machine oil, or a solution of lanoline in white spirit (to D.T.D. 121C) have been used with equal success in this work; preference has been given to the latter, however, since it is normal practice to send out semi-finished aluminium alloy components coated with a protective film of lanoline. The only objection to these surface treatments is that they make the handling of the material difficult.

The Routine Inspection of Rectangular Sections

The crystal wedges used were those supplied with the apparatus. The contact surface of these wedges was flat and the crystal was mounted on a plane making an angle of about 30° with the contact surface. A film of oil transmits the supersonic energy from the crystal to the wedge.

Fig. 3 shows, diagramatically, the manner in which the equipment was assembled. The rectangular spar sections were of 30 ft. length and were carried forward in front of the apparatus on a line of rollers one of which was driven by a geared-down motor. The operator stood opposite the apparatus and held the two crystal wedges on to the surface of the section.

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Fig. 8a.—Photograph of crack shown in Fig. 6a. × 100. Approximate width of crack (across section) 0.045 in.

The supply of lanoline was obtained from a small reservoir and was fed on to the surface, before the crystal wedges, through a drip-feed. After the crystal wedges the section passed under a felt pad which was sprung on to its surface and served to wipe away excess lanoline into a tank placed underneath.

When an indication of a defect appeared, the motion of the section was stopped and the portion containing the defect was explored thoroughly. The rate at which the section could move forward depended to a large extent on the skill of the operator, but an average rate approximated to about 5 feet per minute.

Routine Inspection of Cylindrical Bar

It was thought that in this case it would be an advantage to make the wedge contact faces of the same curvature as the surface of the cylindrical bar. This was tried and, as will be seen later, it is doubtful whether this type of wedge is any better than the type used for rectangular sections.

Fig. 4 shows, diagrammatically, the construction of the carriage upon which cylindrical bars were placed for routine inspection. The bar rested, in a horizontal position, on four rollers and was slowly rotated by driving the shaft carrying one pair of rollers.. The carriage was mounted on wheels and was slowly pushed in front of the supersonic apparatus.

Laboratory Investigation

1. *Rectangular Sections.* Fig. 5 shows a typical indication of the bottom echo from a rectangular section as seen on the cathode-ray tube.

Figs. 6(*a*) and 7(*a*) are photographs of small cracks (marked by circles) in slices cut from a rectangular section (in B.S.S. A2L.40) and Figs. 6(*b*) and 7(*b*) are diagrams of the cathode-ray trace which indicated their presence. Inspection took place from the face AB (see Figs.); it will be observed that the ratio of the depth of the crack, in each case, to the total depth of the section corresponds to the ratio of the distances OX and OY on the tube trace (X being the position of the crack echo and Y the position of the bottom echo). This relationship always holds and it follows that the position of a defect can be estimated fairly accurately. In addition, information regarding the length and width of a defect can be gained by noting the distance, along the section and across it respectively, through which the crystal wedges must be moved before the indication is lost.

Micro-specimens were taken from the slices illustrated in Figs. 6(*a*) and 7(*a*) to show the defects more clearly. Photomicrographs of the cracks are shown in Figs. 8(*a*) and 8(*b*). The widths of these cracks across the section were approximately 0·045 in. and 0·035 in. respectively.

The sections of spars illustrated in Figs. 6(*a*) and 7(*a*) showed a number of fine subsurface discontinuities

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Fig. 8b.—Photograph of crack shown in Fig. 7a. × 100. Approximate width of crack (across section), 0.035 in.

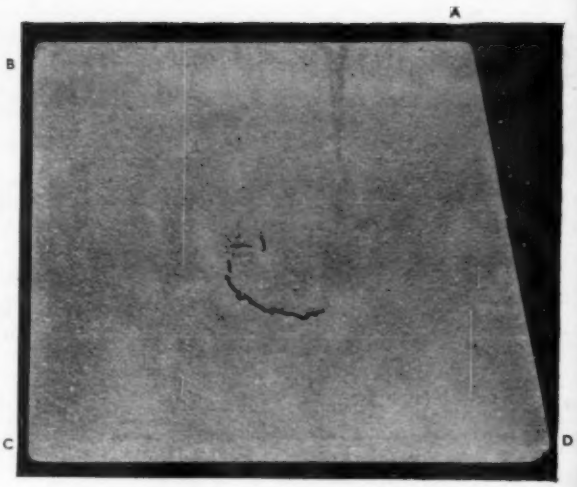

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Fig. 9a.—Large crack in B.S.S. A 2L. 40 spar. ×1

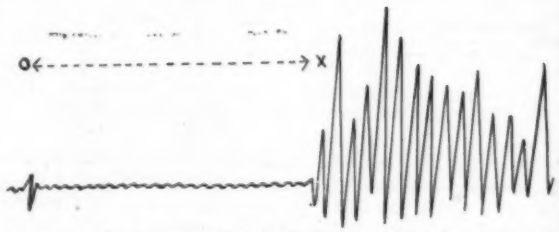

Fig. 9b.—Trace indicating presence of crack shown in. Fig. 9a.

round the periphery (see Figs.) which were of the same order of magnitude as the cracks marked by circles and which were situated within 0·1 in. of the surface. These discontinuities were not detected because any reflection from them was masked by the bottom echo.

In Fig. 9 (*a* and *b*), which shows a large crack, it will be noticed that the high frequency vibration, on the tube trace, moved bodily towards the origin (i.e., the point O on the diagram) and the position of the bottom echo cannot be distinguished. Experience has shown that this type of indication is typical for a large crack

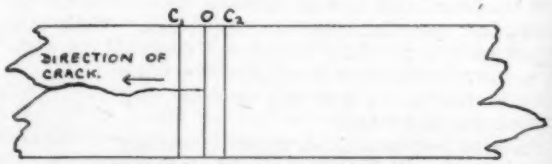

Fig. 10a.—Showing method of sectioning spar to determine the position of the origin of the crack.

and, comparing it with the diagrams of Figs. 6(*b*) and 7(*b*), it is obvious that differentiation between a large defect and a small defect, in this particular type of section, can be made.

Fig. 10 has been included to demonstrate further the accuracy with which the position of a defect can be determined. The crack shown in Fig. 9(*a*) was traced along the

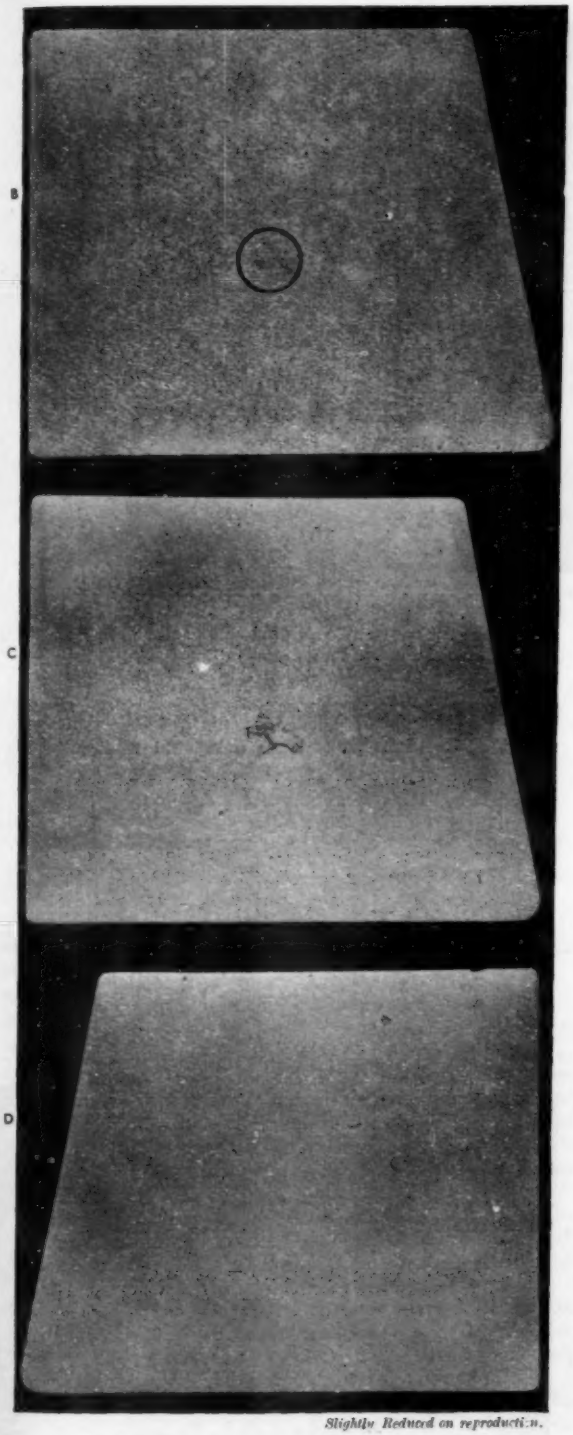

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Fig. 10b.—Showing estimated position of origin of crack. Note crack is just visible. ×1.

Fig. 10c.—Showing development of crack at a distance ¾ in. away from its origin. × 1.

Fig. 10d.—Showing that the crack had disappeared at a distance ¾ in. before the estimated origin.

spar to its origin; here three cross-cuts were made (see Fig. 10(*a*)), one through the estimated position of the origin of the crack and one on either side of it so that two slices, about ¾ in. thick, were obtained. Fig. 10(*b*) shows the etched cross-section through the origin O and the crack is just visible. Fig. 10(*c*) shows the etched cross-section through the cut C_1 and the crack is quite large. Fig. 10(*d*) shows the etched cross-section through the cut C_2 and there is no sign of the crack. Further cutting up of the spar showed that the knowledge of the configuration and position of the crack (evidence of which is not given here) was exceedingly accurate.

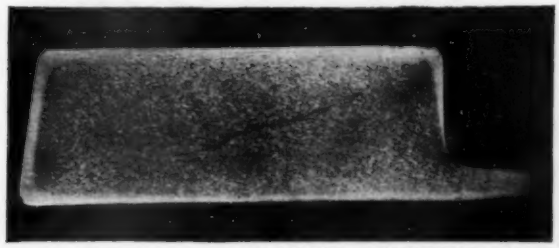

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Fig. 11a.—Large crack in D.T.D. 364A spar—acid etched.

Fig. 11b.—Trace indicating presence of crack shown in Fig. 11a.

Fig. 11 (*a* and *b*) is a photograph of a crack found in a D.T.D.364 spar. Again, Fig. 11(*b*), like Fig. 9(*b*), is indicative of a large crack. This section was examined in a like manner to the previous section of Fig. 9 and an equally accurate determination of the configuration and position of the crack was found to have been made.

2. *Cylindrical Bar.* Nearly the whole of the work on cylindrical bar was carried out on specimens of 6 in. maximum diameter and, as mentioned before, two types of crystal wedge were used; the plane type and the curved type.

With both types of wedge it was difficult to maintain good contact with the surface of the specimen as it rotated and moved forward. The following are the main causes of this difficulty:—

1. Since the surface of the specimen is curved, the liquid transmitting medium quickly runs off and a good film cannot be maintained.

2. In the case of the use of plane wedges, the contact surface is tangential to the specimen surface and only a narrow band of contact is made.

3. In the case of the use of the curved wedges, continual good contact is difficult to maintain because the shoes rarely fit snugly.

In addition to these difficulties the present arrangement suffers from the disadvantage that only a small proportion of the available energy reaches the receiver. Fig. 12 illustrates this. Fig. 12(*a*) shows the plane wedge

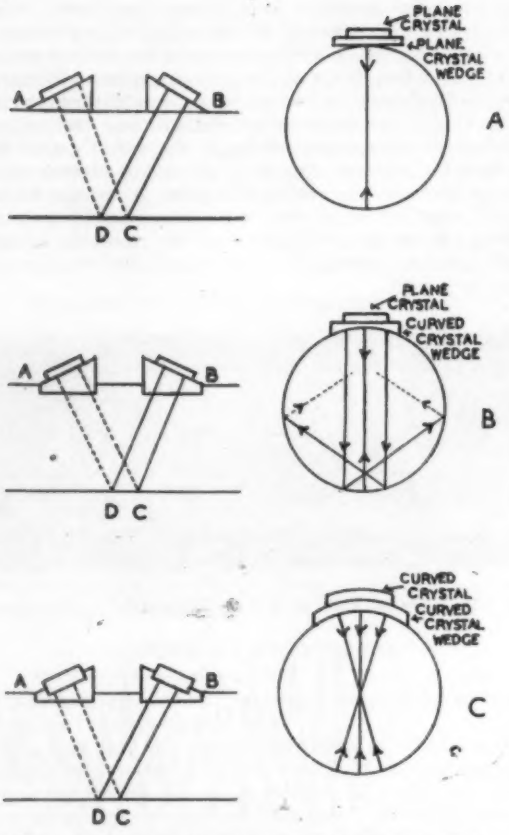

Fig. 12.—Propagation of supersonic energy within a cylindrical bar.

tangent to the surface along the line AB. Only along this narrow band of contact is energy transmitted to the specimen and is reflected back to the receiver at the line CD in a plane through CD normal to the surface. In the case of the curved wedges the energy, emanating from the wedge contact area, is scattered in a manner shown in Fig. 12(*b*) and, if the transmitter is kept fixed while the receiver is moved round the surface, echoes are picked up all the way round. The presence of a defect in a cylindrical bar is indicated but the signal is relatively weak and the interpretation of the tube trace may lead to erroneous conclusions concerning the magnitude, position and configuration of the defect.

It seems that the supersonic beam should be transmitted to the specimen in a radial direction from the whole area of the wedge contact surface. This could be achieved by having curved crystals, in which case all transmitted energy would be in planes normal to the reflecting surface and would be reflected back to the receiver as shown in Fig. 12(*c*).

Fig. 13(*a*) shows a photograph of a crack found in a cylindrical bar of $4\frac{3}{4}$ in. diameter stock (in B.S.S. 2 L.42). Figs. 13(*b*) and 13(*c*) are the tube traces which indicated the presence of this crack when plane wedges and curved wedges were used. From these diagrams it is seen that the indications obtained from the two types of wedge are similar, and little advantage seems to have been gained by using the wedges with curved contact surfaces.

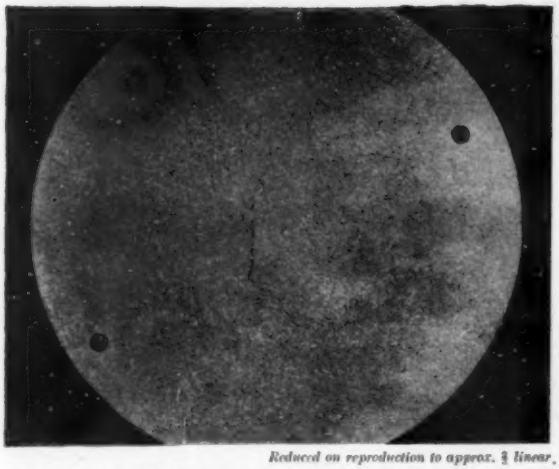

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Fig. 13a.—Central crack and preferential etching in B.S.S. 2L.42 material. ×1.

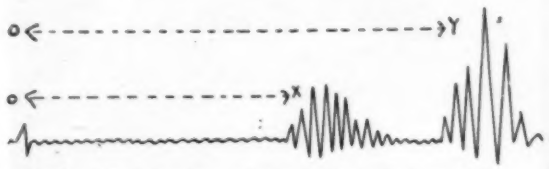

Fig. 13b.—Pattern of trace using flat wedges.

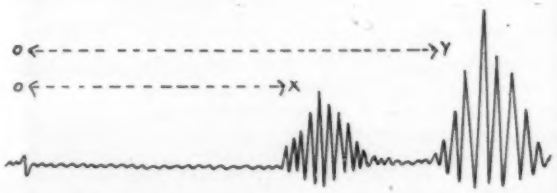

Fig. 13c.—Pattern of trace using curved wedges.

Had the crack in Fig. 13(*a*) been present in a rectangular section its dimensions were such that an indication similar to Figs. 9(*b*) and 11(*b*) would have been obtained, that is, the high frequency vibrations would have moved bodily towards the origin O and the position of the bottom echo would not have been distinguishable. It follows, from the fact that this type of indication is not obtained in the case of cylindrical bar, that it is difficult to differentiate between large and small defects from observations of the tube indication.

Referring again to Fig. 13(*a*) it will be noticed that the crack forms part of a ring of small holes. The interpretation of the tube trace did not predict the presence of such a defect.

Fig. 14 shows a large crack and Fig. 15(*a*) is a diagram of this crack. Around the periphery of the bar [(Fig. 15(*a*)] are marked positions A, B, C, D, E, F, G, H, J and K. The crystal wedges were placed in each of these positions and the corresponding tube trace drawn in each case, as shown in Fig. 15(*b*). Reference to these diagrams further exemplifies the small dimensions of the indication and the difficulty of estimating the position and configuration of the defect.

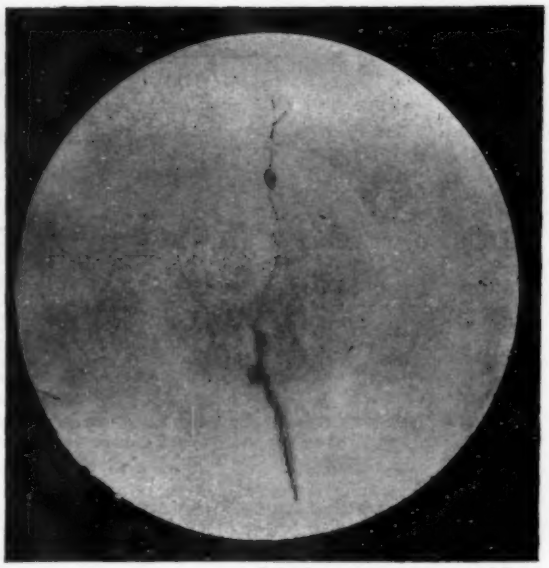

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Fig. 14. Large crack in 5½ in. diameter bar. ×¾

A sample of 5¾ in. diameter extruded stock in an experimental high temperature alloy was examined. The defect shown in Fig. 16(*a*) was found by the indication given in Fig. 16(*b*). Figs. 17 and 18 show photomicrographs of the defect which consisted of a segregation of primary crystals of the $MnAl_6$ type. It will be noted that a number of small cavities are present in Fig. 18, of the same order of size as the primary crystals; in all cases these occurred inside the primary crystals and not

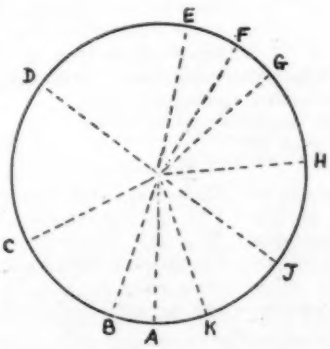

Fig. 15a.—Diagram of crack shown in Fig. 14.

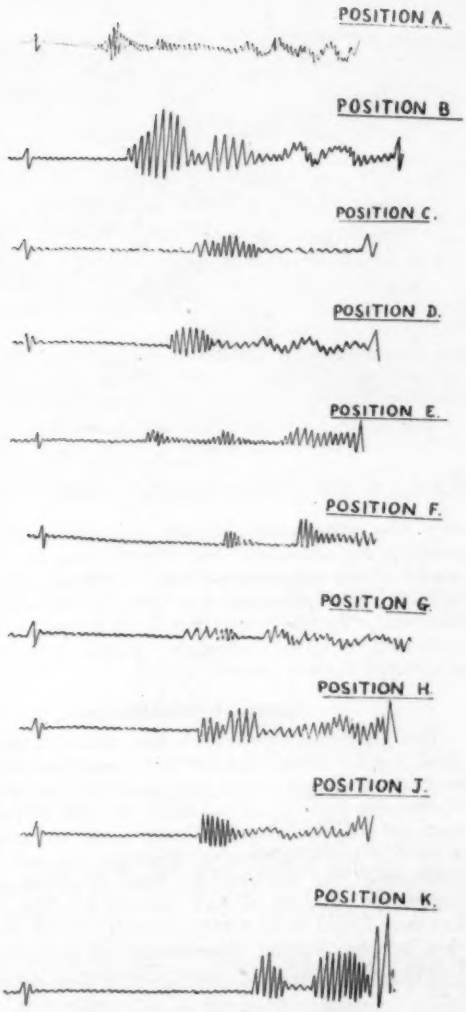

Fig. 15b.—Trace of crack taken at various positions on the bar as indicated in the diagram Fig. 15a.

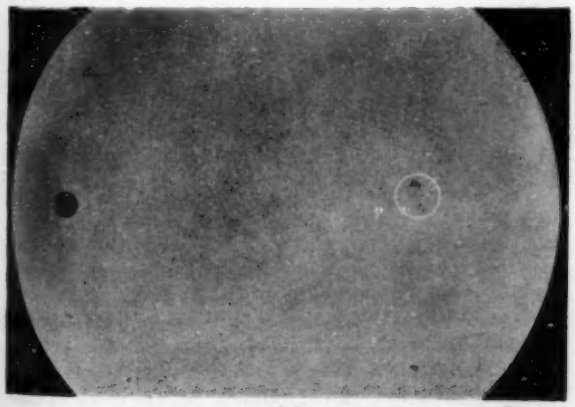

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Fig. 16a.—Small patch of segregate, indicated by circle, in 5¾ in. diameter experimental high temperature alloy. ×1.

in the matrix. The cause of the indication in the trace on the cathode-ray tube is not quite clear; primary $MnAl_6$ type crystals are sometimes hollow in form and such discontinuities would be found by the apparatus but, since the material was extruded, it would be surprising if the hollow crystals had not been broken up and filled in. It is possible that the small cavities were actually formed during the polishing process, in which

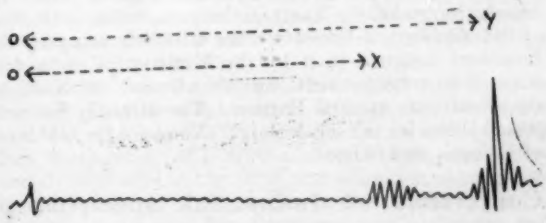

Fig. 16b.—Trace indicating presence of segregate shown in Fig. 16a.

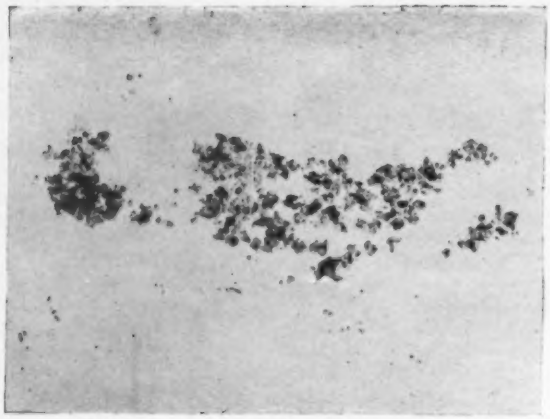

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Fig. 17.–Appearance of segregate shown in Fig. 16. ×10.

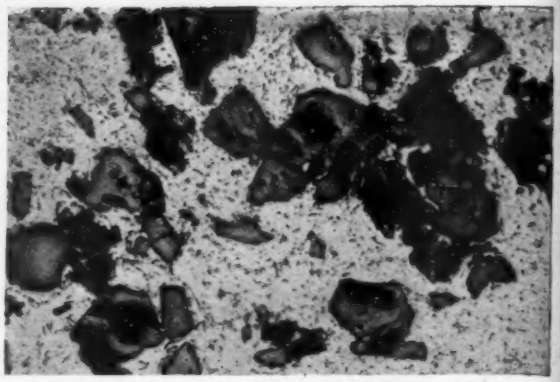

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Fig. 18.—Microstructure of segregate, showing primary crystals of the $MnAl_6$ type containing cavities and cracks. ×100.

case the apparatus detected segregate alone, most possibly by obtaining reflections from the very fine cracks which are always found in primary $MnAl_6$ type crystals. Further experience has confirmed this suggestion since the apparatus has been found incapable of detecting eutectic segregate which was free from associated discontinuities.

General Conclusions

The equipment described has achieved remarkably good results when used for the inspection of extruded sections of simple shape and later models of the same equipment have been modified so that larger sizes of material may be inspected with equal success.

In this investigation an attempt has been made to emphasise the capabilities and possibilities of the supersonic method of flaw detection. The apparatus has been found to be a very powerful tool in its application to the routine examination of aluminium alloy billets and extruded sections. The use of the apparatus not only ensures that sound material is dispatched, but experience has shown that by employing it for the examination of cast billets, it also prevents expensive working and heat-treatment operations being given to inherently defective material. As a result of its use in routine production procedure, the general quality of all material has improved since billets may be examined soon after they are cast and their quality rapidly reported to the persons concerned, thus giving an immediate stimulus to make any necessary changes.

Acknowledgments

The authors wish to express their thanks to Mr. P. W. Rolleston, Vice-President of Aluminium Laboratories Ltd., for permission to publish this work, and to Mr. G. H. Field, Director of Research Division, Banbury, for his encouragement and interest; they also wish to thank their colleague, Dr. H. K. Hardy, for his guidance with the metallurgical examination of specimens.

The Institute of Physics

At the recent annual general meeting of the Institute of Physics, the following were elected to take office on 1st October, 1946: *President:* Prof. A. M. Tyndall; *Vice-Presidents:* Prof. J. A. Crowther, Dr. H. Lowery and Mr. A. J. Philpot; *Honorary Treasurer:* Mr. E. R. Davies; *Honorary Secretary:* Dr. B. P. Dudding; *New Ordinary Members of the Board:* Dr. A. McCance, Mr. W. S. Vernon, Dr. F. A. Vick. Dr. G. F. C. Searle, F.R.S., was elected an Honorary Fellow of the Institute.

The Report for 1945 of this Institute shows that its membership increased by 307, the highest since its foundation in 1918. Early in the year, at the invitation of the Ministry of Education the Institute accepted the functions assigned to it by the Ministry in connection with the arrangements for the award of National Certificates in Applied Physics. The Ministry has now issued Rules for the scheme and "Notes for the Guidance of Colleges and Schools."

Conference on Industrial Spectroscopy

The Institute of Physics announces that its newly formed Industrial Spectroscopic Group will hold its first Conference on Friday and Saturday, 5th and 6th July, in the Auditorium of the Wellcome Research Foundation, 183, Euston Road, London, N.W.1. The Friday meetings will last from 2-30 p.m. to 7-0 p.m. and will be devoted to infra-red absorption spectroscopy. The speakers will be Dr. H. W. Thompson, of St. John's College, Oxford, who will explain the significance of absorption spectroscopy for analytical purposes, and Mr. G. F. Lothian, of the South-West Essex Technical College who will discuss equipment for infra-red absorption measurements.

The Saturday meetings will last from 10-0 a.m. to 5-0 p.m. and will deal only with emission spectroscopy. In the morning Mr. Nickelson, of Woolwich Arsenal, will speak on recent progress in equipment for spectrographic analysis, and Mr. A. Walsh, of the British Non-Ferrous Metals Research Association, will speak on Spectroscopic light-sources for the analysis of metals and alloys. This will be followed by a short discussion and a lunch. In the afternoon Mr. H. T. Shirley, of the Brown-Firth Research Laboratories will read a paper on a statistical examination of sources of error in the spectrographic analysis of alloy steel.

The Hardenability of Steel

Recent American investigations on the hardenability of steel have dealt with such aspects of the subject as the relationship between the hardenability and the percentage of martensite in some low alloy steels, the influence of titanium on the hardenability of steel and the determination of the most efficient alloy combinations for hardenability. A review of these various researches is given in this article.

IT is now generally conceded that if a steel is to develop optimum physical properties in the conventional quenched and tempered condition, the microstructure after quenching should consist wholly of martensite, and that the transformation of austenite to intermediate-temperature products during quenching will have a deleterious effect upon the physical properties particularly the toughness of the tempered material. The criterion of hardenability, however, that at present is most widely accepted and used is based on a microstructure of 50% martensite. In plain carbon steels or in steels of low hardenability, the difference in hardenability expressed in terms of a 50% martensite microstructure and one of full martensite will not be large, and a hardenability only slightly in excess of that indicated by the 50% martensite criterion will suffice to ensure full hardening to martensite. With steels of higher hardenability, however, the difference may be considerable, and in order to determine such relationship an investigation was carried out by J. M. Hodge and M. A. Orehoski[1] with steels of medium hardenability.

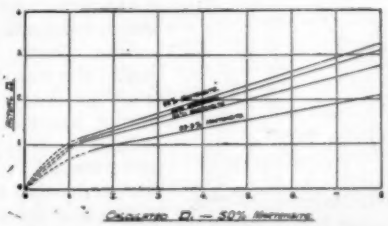

Fig. 1.—Average relationships between hardenability based on higher percentages of martensite and calculated hardenability.

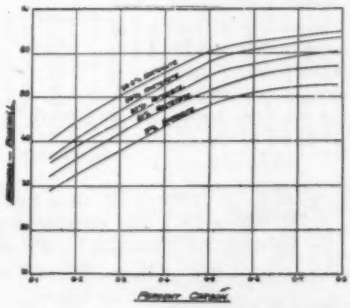

Fig. 2.—Average relationship between carbon content, hardness and percentages of martensite.

The investigation was carried out on plain carbon and alloy steels having carbon contents over the range 0·15–0·79%. The alloys steels contained from 0·5–1·51% nickel, 0·5–1·0% chromium and 0·15–0·39% molybdenum. The experimental procedure consisted of a metallographic study of a series of end-quench tests, to determine the relationship between the percentage of martensite and the distance from the quenched end. The distances were then converted into hardenability values in terms of ideal diameter, and the hardenability value for the microstructure of 50% martensite plotted against that for 99·9%, 95%, 90% and 80% martensite. Rockwell hardness measurements were also made every $\frac{1}{16}$ in. on the surfaces of samples cut from routine Jominy hardenability tests and polished for microexamination, and the results obtained were plotted, and the hardnesses at the distances corresponding to 99·9, 95, 90, 80 and 50% martensite were read from these curves. These hardness values were then plotted against carbon content for each of the microstructures.

The average relationships between hardenability based on higher percentages of martensite and calculated hardenability are given in Fig. 1 and between carbon content hardness and percentage of martensite in Fig. 2. The curves in the first Fig. show that the accuracy of predicting the hardenability in terms of martensite is within approximately 0·3 in. in ideal diameter for low-alloy steels of the types studied, and the carbon-hardness curve in Fig. 2 for 99·9% martensite appears to be accurate within about 3 points Rockwell C.

Grossmann's method for calculating the hardenability of steel from the composition and grain size has gained wide acceptance and, when properly used, has been well proved in practical applications both for estimating the hardenability of known compositions and for selecting the compositions to meet known hardenability requirements. The effect on hardenability of each element present in a steel can be represented by a multiplying factor, the value of which depends on the amount of the element present and the degree of effectiveness of the particular element. As hardenability is one of the most important properties of an alloy steel and one that must be considered in all cases involving heat-treatment, a method has been devised by H. E. Hostetter[2] for the determination of the most efficient alloy combinations for hardenability.

For comparing the alloying elements on standards of cost and hardenability, it is first necessary to select increments of alloy content of equal cost and those calculated on current values for chromium, molybdenum and nickel were found to be 0·11, 0·02 and 0·056%, respectively. In determining the most efficient alloy combination, use is made of what are termed increment factors which are a quantitative means of expressing the general qualitative principle that, in terms of multiplying factors, each added increment of alloying element gives decreasing returns in hardness. The calculated increment factors for chromium, molybdenum, and nickel were determined and are given in Table I, while the most efficient combinations of chromium, molybdenum and nickel for various levels of hardenability are given in Table II.

To determine the composition which will have an average D_1 (the ideal critical diameter which is the criterion of hardenability) of 4·80 for an average alloy free composition of 0·40% carbon, 0·85% manganese and 0·25% silicon, the D_1 of the alloy-free composition is first calculated and found to be 1·12. The combined multiplying factor for chromium, molybdenum and

1. *Metals Technology*, 1945, Vol. 12, No. 6; A.I.M.M.E. Tech. Publication No. 1800, pp. 1–2.

2. *Metals Technology*, 1945, Vol. 12, No. 6; A.I.M.M.E. Tech. Publication No. 1905, pp. 1–10.

TABLE I.—CALCULATED INCREMENT FACTORS

Chromium			Molybdenum			Nickel		
Cr %	Multiplying Factor	Increment Factor	Mo %	Multiplying Factor	Increment Factor	Ni %	Multiplying Factor	Increment Factor
0·00	1·00	—	0·00	1·00	—	0·00	1·00	—
0·11	1·26	1·26	0·02	1·062	1·062	0·056	1·04	1·040
0·22	1·52	1·21	0·04	1·125	1·059	0·112	1·08	1·038
0·33	1·78	1·17	0·06	1·187	1·056	0·168	1·12	1·037
0·44	1·96	1·101	0·08	1·25	1·053	0·224	1·16	1·036
0·55	2·06	1·051	0·10	1·312	1·050	0·28	1·20	1·034
0·66	2·155	1·046	0·12	1·375	1·048	0·336	1·24	1·033
0·77	2·245	1·042	0·14	1·437	1·046	0·392	1·28	1·032
0·88	2·33	1·038	0·16	1·50	1·044	0·448	1·32	1·031
0·99	2·41	1·034	0·18	1·562	1·042	0·504	1·36	1·030
1·10	2·485	1·031	0·20	1·625	1·040	0·56	1·40	1·029
1·21	2·555	1·028	0·22	1·687	1·038	0·616	1·44	1·029
1·32	2·62	1·028	0·24	1·75	1·037	0·672	1·48	1·028

TABLE II.—MOST EFFICIENT ALLOY COMBINATIONS

Element %			Multiplying Factor			Cr Factor × Mo Factor × Ni Factor
Cr	Mo	Ni	Cr	Mo	Ni	
0·10	0·00	0·00	1·24	1·00	1·00	1·24
0·20	0·00	0·00	1·47	1·00	1·00	1·47
0·30	0·00	0·00	1·71	1·00	1·00	1·71
0·40	0·00	0·00	1·92	1·00	1·00	1·92
0·51	0·02	0·00	2·03	1·06	1·00	2·15
0·53	0·06	0·00	2·04	1·19	1·00	2·43
0·57	0·10	0·00	2·08	1·31	1·00	2·72
0·66	0·14	0·00	2·16	1·44	1·00	3·11
0·77	0·18	0·00	2·25	1·56	1·00	3·51
0·82	0·20	0·06	2·29	1·63	1·04	3·88
0·88	0·22	0·13	2·33	1·69	1·09	4·30
0·91	0·24	0·17	2·36	1·75	1·12	4·62

nickel to provide a D_1 of 4·80 is the term (Cr × Mo × Ni) where Cr is the multiplying factor for chromium, Mo for molybdenum and Ni for nickel.

$$(\text{Cr} \times \text{Mo} \times \text{Ni})\,(1{\cdot}12) = 4{\cdot}80$$
$$\therefore (\text{Cr} \times \text{Mo} \times \text{Ni}) = 4{\cdot}28$$

which is equal to 0·88% chromium, 0·22% molybdenum and 0·13% nickel. (Table II).

As a certain amount of disagreement has arisen on the effect of titanium on the hardenability of steel, an investigation has been carried out by G. F. Comstock[3] on three titanium steels, containing 0·055, 0·36 and 0·777% of titanium, and two non-titanium steels heat-treated in the same way and between other titanium steels heat-treated in several different ways. Hardenability test specimens of 1 in. dia. were machined from the five steels, after treatment for 4 hours at 1,200° C. followed by air cooling, and after heating in various ways at 1,200°, 1,080° and 900° C., were end-quenched by the standard Jominy procedure. The depth of hardening was estimated not only from the microstructure but also from the hardenability curve.

The results obtained showed that the effect of titanium on the hardenability of steel depended on the heat-treatment of the steel, on whether total titanium or acid-soluble titanium is used as a basis of calculations and on the amount of titanium present relative to the carbon, so that a single definite value cannot be given for titanium as has been done for other common alloying elements in steel, without careful qualification as to the correct conditions under which it may be applied. In general, however, small amounts of titanium below 0·1% increase the hardenability quite effectively, while above 0·2% the effect is reduced in proportion to the titanium content, until, with sufficient titanium, steel may lose its hardenability completely.

3. *Metals Technology*, 1945, Vol. 12, No. 6; A.I.M.M.E. Tech. Publication No. 1904, pp. 1–15.

In steels of the usual forging grade with about 0·20–0·5% carbon and heat-treated at the usual temperatures, titanium contents up to about 0·2% increase the hardenability, the multiplying factor being equal to one plus about three times the total titanium content, which places small amounts of titanium in about the same class with phosphorus, chromium and molybdenum for promoting hardenability. Larger amounts of titanium decrease the hardenability, however, by the formation of an insoluble carbide, and above about 0·6% total titanium the factors become less than unity, indicating a softening effect. With steels containing more than about 1·4% manganese the factor for total titanium may be higher than with ordinary manganese contents, or one plus about five times the titanium content up to a maximum at about 0·1% titanium and decreasing to below unity with less than 0·6% titanium depending on the carbon content of the steel.

The investigation also showed that the hardenability factors for titanium can be based on acid-soluble instead of total titanium, provided that account is taken also of the fact that the insoluble titanium decreases hardenability by rendering ineffective a proportional amount of carbon. With reasonable fine-grained steels of forging grade, the hardenability factor for soluble titanium up to about 0·02% is equal to one plus about 33 times the soluble titanium percentage which represents a stronger influence on hardenability than has been found for any other element except boron. In most steels, however, the decrease in hardenability due to the carbide-forming action is of far greater importance.

Comparing the hardening characteristics of forging steels of about 0·37 to 0·40% carbon, with moderate manganese and silicon, and negligible contents of other alloys except vanadium and titanium some interesting relationships were obtained. Vanadium steel was found to have a greater hardenability than titanium steel, and low titanium steel greater than high titanium steel when hardened either at 900°, 1,080° or 1,200° C. When quenched from 1,200° C., the depth of hardening compared with that of coarse-grained plain carbon steel was increased by vanadium or up to 0·36% titanium, but decreased by 0·78% titanium; when quenched from 1,080° C., the depth of hardening was increased by vanadium or 0·055% titanium, but decreased by 0·36% or more titanium; and when quenched 900° C. the depth of hardening was decreased by both vanadium and titanium. The vanadium steel was always coarser grained than the higher titanium steels and even coarser than the low-titanium steel after heating to 1,080° or 1,200° C. The increased hardening effect due to titanium was found after air-cooling as well as quenching, but was mostly lost after only 20 minutes at 900° C., although some traces of it persisted for more than an hour. This was not true of the vanadium steel, in which, however, some evidence of slight secondary hardening appeared with the slower rates of cooling.

Argentine Naval Contract for Britain

THE TUBE INVESTMENTS GROUP of engineering industries announce that an order has been received for 73,219 cold-drawn steel boiler tubes for the Argentine navy.

Work on the contract is now being undertaken at their Tubes Limited works, at Aston, Birmingham, which produced over five and a half million boiler tubes for the Allied navies during the war.

Metallurgy and the Development of the High-speed Diesel Engine

By C. C. Hodgson
Chief Metallurgist, Leyland Motors, Ltd.

In the manufacture of the compression ignition engine, as in most phases of the automobile industry, design and manufacture are closely linked with metallurgy. Some of its associated metallurgical problems were discussed by the author at a recent meeting of the Sheffield Metallurgical Association. He began by giving a brief description of the type of engine under consideration and its characteristics and subsequently outlined the materials used and elaborated a few of the more interesting aspects. By courtesy of the author and of the Association we are able to present this contribution on a subject in which performance and cost are primary considerations.

THE compression ignition engine, also known as the Diesel engine (although purists maintain that Akroyd Stuart had more right to the title than Dr. Diesel), or simply as an oil engine, first attracted serious interest as a power unit for commercial vehicles and buses about 20 years ago, and it is with its use in this connection that this discussion is concerned. Although a comparable petrol engine would be considered a large one, this type of Diesel engine is small as Diesel engines go, and although it is usually referred to as a small, high-speed engine, in actual fact its maximum governed speed (which usually lies somewhere between 1,500 and 2,000 r.p.m., according to type), is less than that of a petrol engine of similar horse power.

The characteristic of the Diesel engine that has most commended it to commercial vehicle users is its fuel economy. It can convert more of the potential heat value of the fuel into useful work than its competitor, the petrol engine, and under average conditions a compression ignition engined vehicle, competing with a petrol engine of similar power, can be expected to travel at least half as far again on a gallon of fuel. Development of the road Diesel engine has, therefore, been most rapid in countries where fuel is expensive, either from intrinsic cost or from heavy taxation.

The petrol engine draws into its cylinders a combustible mixture of air and petrol vapour, or mist, and the charge is fired by the sparking plug as the piston nears the top of the compression stroke. The Diesel engine, on the other hand, inhales a charge of pure air during the suction stroke, the fuel oil is injected in the form of a fine spray as the piston nears the top of the compression stroke and is ignited by the high temperature of the compressed air. In order to produce a temperature high enough to ensure ignition the air has to be highly compressed, a compression ratio around 15 to 1 is usual, compared with, say, 5½ to 1 for a petrol engine. These differences influence the pressures developed during combustion; under full load the maximum pressure in the Diesel engine will be roughly half as much again to twice that in a petrol engine of similar power, and furthermore, whilst the maximum pressure in the petrol engine varies roughly in proportion to the percentage of full load, the maximum pressure in the Diesel engine is still quite high even when idling. (See Fig. 1).

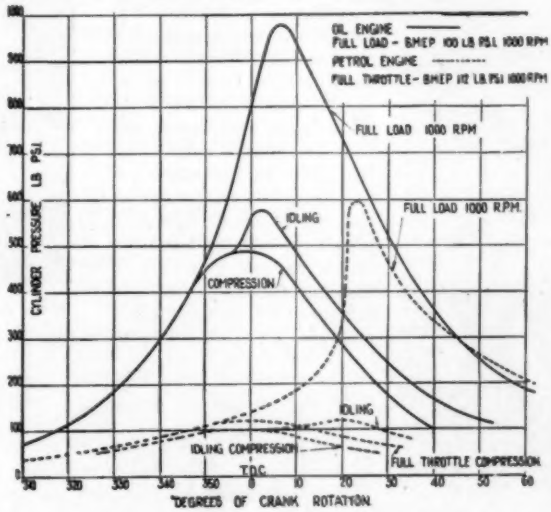

Fig. 1.—Gas pressures on oil and petrol engines.

This difference is of the utmost importance both to the designer and metallurgist because it necessitates an increase in size and strength of most of the major components, from engine castings to crankshafts. A good idea of what these alterations amount to can be gained from the comparison between the size of crankshafts (Fig. 2) and connecting rod drop forgings (Fig. 3) used for commercial vehicle petrol and oil engines of similar power output.

The designer of these small Diesel engines started with the accumulated experience of a quarter of a century in the use of materials for building petrol engines. Some components had to be increased in size in order to obtain the required rigidity, and materials similar to those already used for petrol engines were amply strong for these. In some other respects the problems of the aero engine and Diesel engine were almost identical, and their solution has been along similar lines. The materials used include, on the ferrous side, several varieties of high quality cast iron, and the usual heat-treated alloy steels, as well as those suitable for case-hardening and nitriding, and heat-resisting steels; and in the non-ferrous group, light alloys, copper and zinc base alloys, and complex hard alloys and surfacing materials. It is proposed to deal with the subject of materials by sketching the general picture in outline and elaborating only a few of the more interesting aspects.

Iron Castings

The amount of cast iron used in a small, high-speed compression ignition engine is considerable. Iron castings

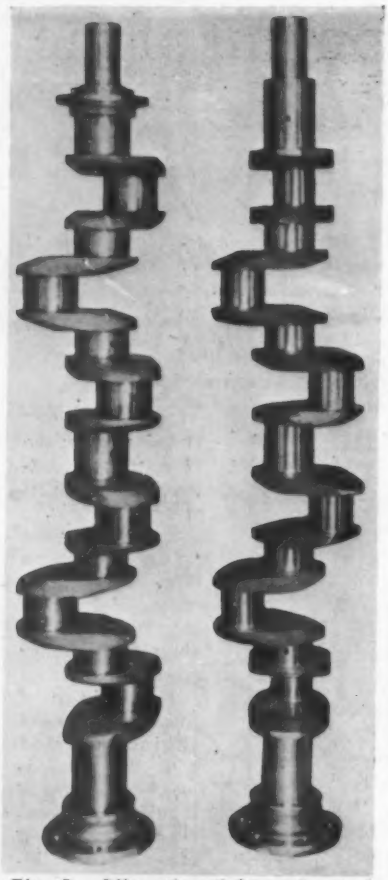

Fig. 2.—Oil engine (left) and petrol engine crankshafts for engines of similar horse-power.

are generally used for cylinder head and block, and for the crankcase. The iron used is of the high-grade cylinder type usually lightly alloyed. If cylinder liners are fitted, these are also usually made of cast iron, and this form of construction permits the use of different irons for the two purposes and makes it possible to select compositions that will give strong, sound castings for the block, and material of high wear resistance for the liners, instead of having to attempt a compromise.

It was usual, before 1940, to cast crankcases in light alloys in order to save weight. The shortage of light alloys early in the war made it necessary to use cast iron as a substitute. This change, whilst it increased slightly the weight of the engine and increased the time required for machining, was, on the whole, an improvement; it gave a more rigid assembly and removed the troubles caused by differences in expansion between the cast iron block and the light alloy crankcase.

Apart from temporary distortion caused by differences in coefficients of thermal expansion, it was not perhaps generally realised how much permanent movement could take place in certain light alloys when moderately heated and stressed. The permanent distortion of a few light alloys, determined at a temperature of 100° C. on specimens machined from sand cast bars 1 in. diam. and subjected to a constant compression load of 10 t.s.i. for periods up to eight hours, is illustrated in Fig. 4. Cast iron and some of the fully heat-treated light alloys tested under similar conditions take no measurable permanent set.

It is not contended that changes of such a magnitude take place in the engine; the conditions of test were purposely exaggerated in order to get comparable results quickly, and in common with all accelerated tests such results must be regarded with a certain amount of suspicion and accepted only with reserve. But before returning to the use of light alloys for such parts, designers will no doubt take these and other factors into consideration.

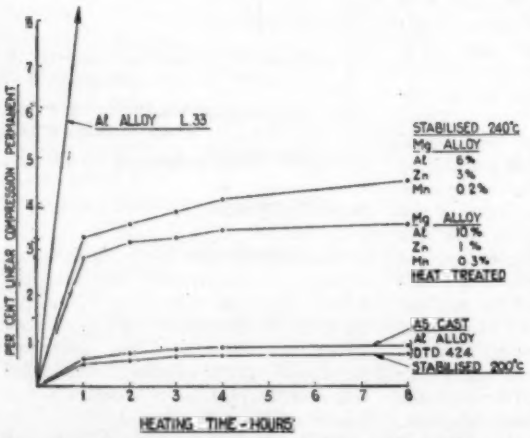

Fig. 4.—Effect of prolonged loading under compressive stress at elevated temperature on some cast light alloys.

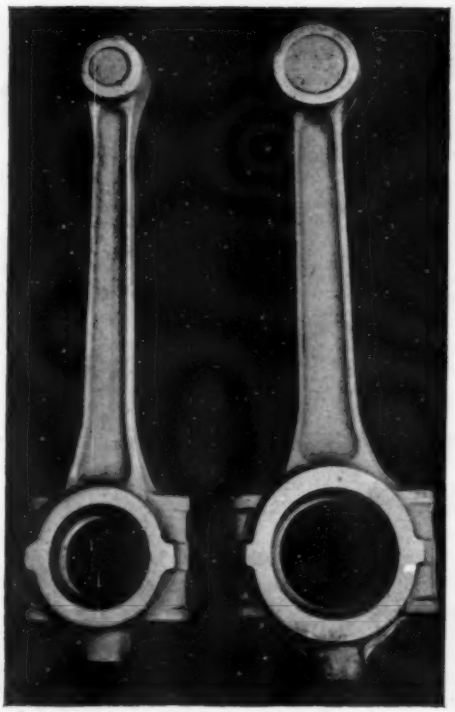

Fig. 3.—Petrol engine (left) and oil engine connecting rod drop forgings.

Pistons

The mention of light alloys leads naturally to a consideration of piston alloys. One of the most popular of these has been "Y" alloy, which was developed largely as a result of experiments on piston alloys during the 1914–18 war. Particulars of this alloy were published in the 11th report of the Alloys Research Committee of the Institution of Mechanical Engineers, the last report on a series of researches by distinguished British metallurgists, commenced by Roberts-Austen with his report in 1891 and continued by Gowland, Carpenter, Hadfield, Longmuir, Edwards, Rosenhain, Lantsbury, Archbutt, and Hanson. The alloy is still popular; newer alloys have been developed, usually rather complex ones, but they can be regarded as being based originally on "Y" alloy, except for the high-silicon, low-expansion alloy of the type covered by specification DTD.324.

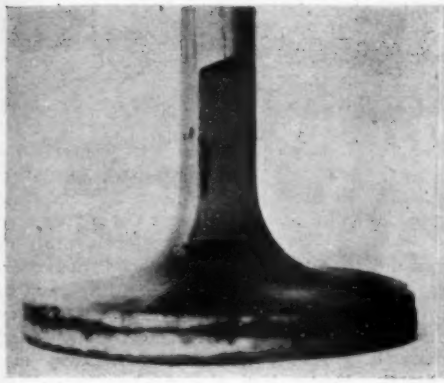

Fig. 5.—Sectioned valve forging showing annular region of large crystals developed more in the interior than at the surface.

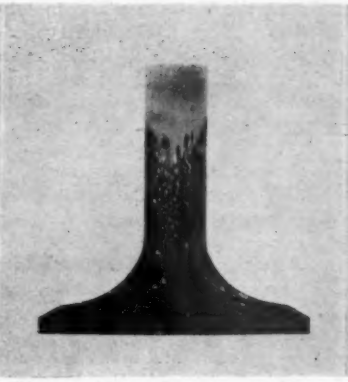

Fig. 6.—Photomacrograph of section of valve head and neck showing development of large crystal grains produced by heating following critical strain.

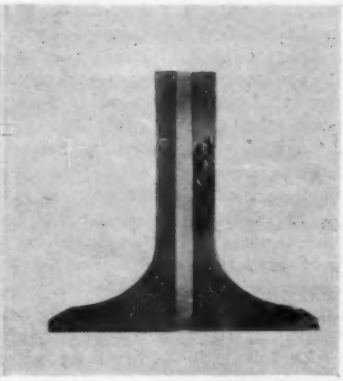

Fig. 7.—Photomacrographic sections of portions of the valve shown in Fig. 6 after re-heating and quenching from 950° C. (right), and 1050° C.

Both cast and wrought pistons are in use and the question which is the better naturally arises, the fact is that each is performing satisfactorily in the type of engine under discussion. In order to produce a forged or pressed piston an alloy possessing considerable plasticity when hot must be used, and it might be argued, therefore, that it should be possible to evolve a more suitable type of alloy capable of being cast but without the necessary plasticity when hot to permit of its being mechanically worked. Against this it might be said that it is not usually possible to produce castings that are quite as sound as forgings or pressings.

Ductility is certainly not necessary, after all there is not much wrong with cast iron except its density. What is needed is freedom from dimensional changes, a good fatigue resistance, good wearing properties, high heat conductivity, and low coefficient of expansion (these last two are usually mutually exclusive amongst suitable light alloys), and lightness; and it is the question of weight that has mainly caused the eclipse of cast iron. The answer to the claims of the cast and wrought piston for the small, high-speed compression ignition engine has still to be given.

Valves

Much work had already been done to establish the requirements of a material suitable for the manufacture of petrol engine exhaust valves at the time Aitchison gave his classical paper to the Institution of Automobile Engineers in 1920, and since then new materials have been found and new methods developed to meet the requirements of high performance aero engines. There was to hand all that was needful for the compression ignition engine valve. In most respects this works under much less onerous conditions than the petrol engine exhaust valve. The exhaust gas temperature is lower, the valve is not intermittently washed with a flood of hot gas charged with the corrosive and destructive decomposition products of "dope" from the fuel. But, although operating conditions are less likely to be destructive, the loss of tune that accompanies imperfectly seating valves can be serious in the compression ignition engine with its high compression ratio. The answer is the provision of hard-metal faced seatings in the head and to the valve, and to improve the wear resistance of valve stem and guide. The procedures are generally known and there is not time to enter further into them here nor into the earlier endeavours to overcome the difficulties by more simple means.

The ordinary silchrome steel of the En.52 type is quite suitable for most C.I. engine exhaust valve applications. One would think that there was surely nothing new to say about this steel which is now between 20 and 25 years old, but strangely enough there is. The principal specifications covering it (D.T.D. 13 B., En.52 and V.24) do not exclude the possibility of valves which are in a brittle and coarsely crystalline condition being supplied to the user. Why? Because the hardening temperature specified or the lower limit of the temperature range, where the range is specified, is too low to ensure complete recrystallisation of the steel. Hardened and tempered valves, quenched from 950° C. or the lower end of the range, usually consist of spheroidal carbide in a matrix of chromiferous ferrite, sometimes with a little sorbite, whilst those quenched from 1050° C., or the upper part of the range, have a matrix of sorbite after tempering. The latter will be completely recrystallised; the former will not, they will, in fact, have largely the crystal grain size left after forging if the finishing temperature is above the recalescence point, as is usual with hammer-forged or goffed valves. But if the finishing temperature is low, a condition of critical strain may arise which can lead to a truly alarming amount of crystal grain growth. The extent of the region of large crystals varies widely according to the imposed conditions. It may consist of a thin shell, no more than a few thousandths of an inch deep, extending over the whole or only part or parts of the head and neck, it may be almost imperceptible at the surface but of considerable extent below the surface (see Fig. 5), or it may extend completely through the section (see Fig. 6). The effect of reheating portions of the valve shown in Fig. 6 to temperatures of 950° C. and 1050° C. can be seen from the macro sections depicted in Fig. 7.

There are, of course, many factors which might contribute to the formation of such large crystals: chemical composition, decarburisation, cold work in the bar or rod used, small machining allowances, method used for producing the forging (regarding the word as a generic term which includes extrusion and electrically upsetting), portions of the valve plastically deformed at a temperature below that of complete recrystallisation, and low hardening temperature, or an attempt to use the prior forging temperature as a quenching temperature (i.e., omitting reheating for hardening); but all the facts are

consistent with the hypothesis that the phenomenon is primarily due to a condition of critical strain combined with too low a hardening temperature. From this array of possible factors metallurgists should have no difficulty in visualising a number of different combinations which could be responsible for one or other of the types of the defect. It is clearly a condition to which the engine builder is entitled to direct attention and for which he expects the remedy to be provided. Unfortunately the acceptance tests imposed by the specifications are not of a kind likely to disclose the presence of the condition.

Bearings and Crankshafts

Probably none of the components, common to the petrol engine and the compression ignition engine, have received as much attention and been the subject of so much development work as the connecting rod bearings and the crankshaft. The practice in petrol engine design had become pretty well standardised on a medium carbon steel crankshaft, with sometimes a small amount of alloy added, used in the normalised or quenched and tempered condition, with a Brinell hardness somewhere around 200, and the bearings for both crankshaft and connecting rod were invariably lined with one or other of the tin-base babbitts. This was the position at the time of the sudden expansion in the use of the small compression ignition engine for road transport.

It was natural that this combination of materials should be carried over to the new engines, and for those in the lower speed ranges it continued, and still continues, to give fairly good results. For engines in the higher speed ranges it soon became apparent that the tin-base white metal which had proved so satisfactory for petrol engines was no longer adequate. Bearing life before failure was considerably lower than in the corresponding petrol engine. Now began one of those periods of experiment which are common in engineering, improvements or changes in one direction lead directly to the necessity for alterations in another.

Engine bearing failures, except in comparatively rare examples caused by lack of lubrication, are almost invariably due to fatigue. The fatigue cracks start at the surface and travel inwards through the bearing metal. If this is a white metal the cracks grow until they reach the region of the stronger tin-iron alloy at the white-metal-steel interface when they are deflected and follow a path roughly parallel to the steel back or shell. Ultimately this leads to a piece, or more usually pieces, of the white metal breaking away.

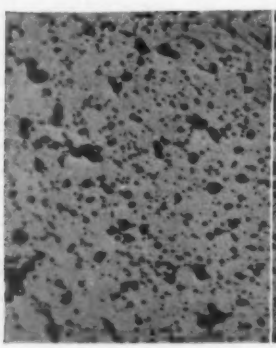

Fig. 8.–Copper-lead alloy: centrifugally cast bearing. × 100.

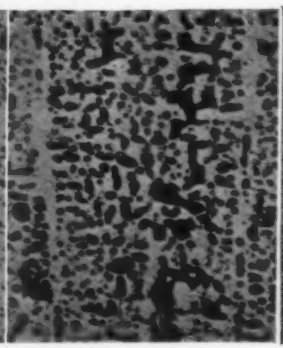

Fig. 9.–Copper-lead alloy: statically cast bearing. × 100.

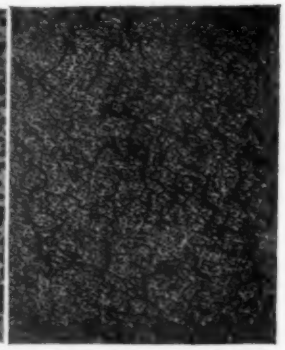

Fig. 10.—Copper-lead alloy: thin strip bearing. × 100.

Bearing Development

There were objections to any attempt to reduce the stresses on the bearings by increasing their length, and attention had to be given to increase their life by finding alloys that had a higher resistance to fatigue. Harder babbits and alloys, of which the metals cadmium, aluminium, silver or copper formed the base, were used, and a list including all the alloys investigated would be quite a long one. Generalising, one may say that as the strength and fatigue resistance of the alloys increased, their tendency to wear the journals and crankpins also increased. In other words, judged from the effect they had on the steel, they were inferior bearing metals.

As a result of experiment and practical experience, the choice has fallen on the copper-lead bearings, containing between 25% and 35% of lead. Copper-lead bearings are stronger than alloys based on metals of lower melting point and consequently much less likely to fail from fatigue. The equipment used for making such bearings has often been very elaborate, and many of the methods used to prevent the separation of lead (to which the alloy is prone, because of the wide interval separating the temperatures of liquids and solids) are very ingenious. The older copper-lead bearings were cast on heavy steel backs but the modern trend for automotive use is to use thin strip bearings. The behaviour of the bearing metal towards the shaft has been improved by a thin surface plating of lead, and the corrosion resistance of the lead has been improved by plating on a flash of indium which is afterwards thermally diffused into the lead.

Photomicrographs of the copper-lead layer of centrifugally cast, statically cast, and strip bearings are shown in Figs. 8, 9 and 10. It is interesting how a remark can date: Gulliver, in his admirable book "Metallic Alloys," 5th edition, published at the end of the 1914-18 war, could write that " the binary series copper-lead has in itself little practical importance. . . . A mixture of equal parts of copper and lead, with the addition of 1% nickel, has been used for the same purpose " (i.e., for bearings). In the more recent war practically all the aircraft engines, on both sides, were fitted with copper-lead bearings, and the enemy relied to a very large extent on similar alloys on the home front as well.

Harder Crankshafts

This work on bearing metals naturally directed attention to the crankshaft. The desired aim of increasing bearing life had been achieved, but only at the expense of increased crankshaft wear, the problem now was to readjust the balance and obtain a resistance to wear at least equal to that of the petrol engine. Heat-treating the steel to a higher Brinell hardness was the immediately obvious reaction, a change that usually meant using a more highly alloyed steel. This alteration, whilst it caused an improvement, was not altogether satisfactory and there followed the use of methods for specially hardening crankpins and

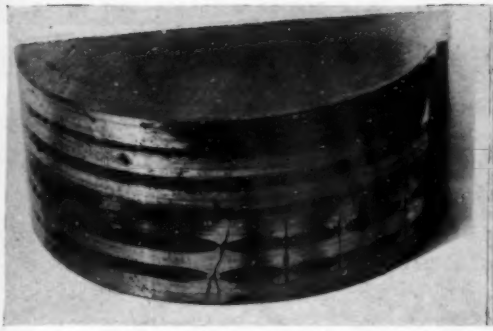

Fig. 12.—Case-hardened crankpin softened and cracked through overheating.

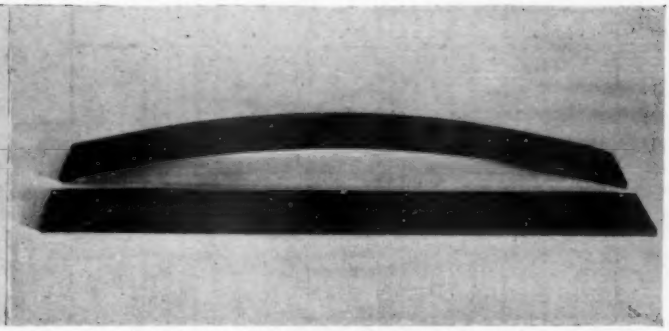

Fig. 13.—Specimens illustrating increase in volume caused by nitriding.

journals such as case-hardening electrolitically deposited chromium, flame and Tocco methods of surface-hardening and nitriding. It would be out of place here to attempt to assess the relative merits or demerits of these methods of producing a hard-surfaced crankshaft, but one may perhaps be permitted a statement of personal opinion and say that a nitrided crankshaft used with copper-lead bearings offers a combination of desirable properties which places this construction in the lead at the present time. Certainly it is no exaggeration to say that the result of all this work, carried out in different places and by different sections of the automobile and aircraft industries, has been to place at the disposal of the commercial vehicle user engines with a combination of bearings and crankshaft which have a rate of wear not only equal to but actually considerably less than that of the petrol engine at the time the investigations started.

Advantages of Nitriding

Nitriding confers on the crankshafts (considered as a component of the bearing combination) a property unique amongst the hard-surfaced crankshafts, it is very resistant to softening when heated, thus local and temporary overheating, insufficient to cause seizure but high enough to cause appreciable softening of a surface layer depending on martensite for its hardness, is entirely without effect on the hardness of the nitrided surface. The effect of reheating on the hardness of the case of a case-hardened Ni-Cr steel and a nitrided Cr-Mo-V steel is illustrated by the curves in Fig. 11, specimens heated for a period of one hour. Tests have been made under conditions that caused severe seizure of aluminium alloy bearings on a nitrided crankshaft, so severe that some of the aluminium alloy appeared to be welded to the crankpins, and yet after the aluminium had been dissolved in caustic soda solution the nitrided surface was found to be quite undamaged. Another danger from the overheating of case-hardened or surface-hardened crankpins or journals is unavoidable: the volume change that accompanies the transformation

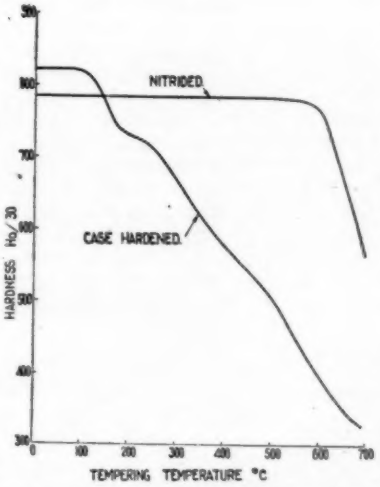

Fig. 11.—Effect of heating for one hour on the surface hardness of case-hardened and nitrided steel.

of martensite into secondary troostite or sorbite is liable to neutralise the compressive stress in the hardened surface layers and may even proceed so far that a tensile stress is produced at the surface, when this occurs cracking is not unlikely. An example of this type of failure is illustrated by Fig. 12, a case-hardened crankpin which has suffered partial seizure with a copper-lead bearing, the specimen has been lightly polished and etched and the difference in colour of the circumferential bands is due to the different amounts of tempering that have taken place during the period of overheating, the cracks formed can also be seen.

High surface hardness and resistance to softening when heated are not the only advantages conferred by nitriding. The importance of surface stress in relation to fatigue failure has long been recognised in a general way. Perhaps the comparatively recent publications of Almen have done more than anything to drive home just how important it is consciously to apply these principles to machine components with the object of improving their resistance to fatigue. Metallurgists as a body should feel grateful to Almen for the unequivocal way in which he has pointed out where the responsibility for much of our broken machinery lies, and for indicating so clearly how much of this waste can be overcome. But to return to nitrided crankshafts: nitriding enables one to produce crankshafts with the right sort of stress in the right place and entirely free from harmful stress. Nitriding causes the nitrided steel to expand, but as in most practical applications the thin, nitrided case is backed-up by a considerable mass of

Fig. 14.—Fatigue fracture of nitrided specimen. × 3½.

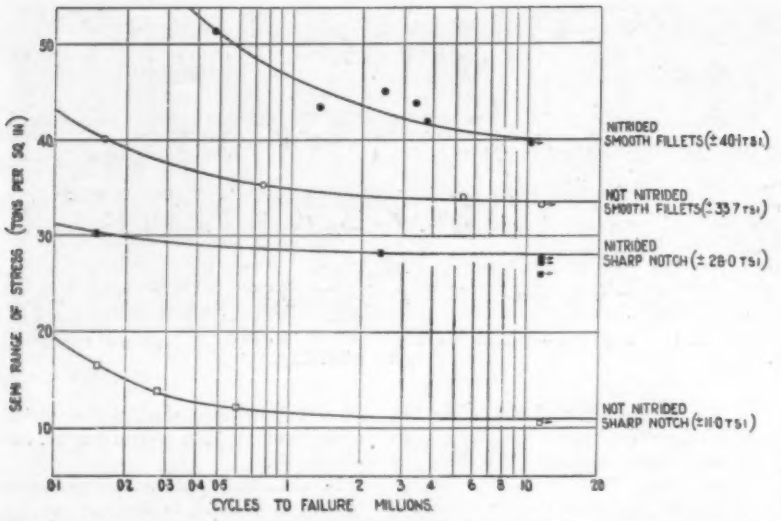
SERV. RANGE OF STRESS (TONS PER SQ. IN.)
50
40
30
20
10
NITRIDED SMOOTH FILLETS (± 40·1 TSI)
NOT NITRIDED SMOOTH FILLETS (± 33·7 TSI)
NITRIDED SHARP NOTCH (± 28·0 TSI)
NOT NITRIDED SHARP NOTCH (± 11·0 TSI)
·01 ·02 ·03 ·04 ·05 1 2 3 4 6 8 10 20
CYCLES TO FAILURE MILLIONS

Fig. 15.—S/N curves illustrating the influence of nitriding on resistance to fatigue.

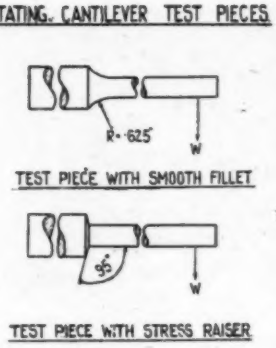

Fig. 16.—Wöhler test pieces used to investigate the effect of notches.

unaltered material, and is thereby restrained, the result is a nitrided part with a compressive stress in the skin. This can be demonstrated quite neatly by nitriding a thin strip of steel on one face only, Fig. 13; the flat strip was nitrided all over, one face of the bowed strip was protected by tinning before nitriding, the expansion of the nitrided skin has caused the strip to bend as shown. So far as a crankshaft is concerned this compressive stress in the nitrided skin gives it a start, so to speak, over the fatigue stress to which it will be subjected in service, actually the locus of failure is removed from the surface to the inside of the forging at or near the inner limit of the nitrided skin, Fig. 14. Another and probably still more important property of the nitrided skin is that it much reduces the evil influence of stress raisers—and all crankshafts are unavoidably full of stress raisers. These properties of the nitrided case are fully confirmed by the endurance curves reproduced in Fig. 15, obtained from Wöhler tests using test pieces having the forms shown in Fig. 16. The results of full scale tests on crankshafts are in entire agreement.

Two Fallacies. It may not be out of place, especially to those interested in the production and use of nitriding steels, to expose two fallacies which have gained a very firm hold, namely: the efficacy of stabilising, and that the hardness of the nitrided case is not affected by preliminary heat-treatment.

He would be a very rash man who would say that stabilising was of no value at all under any circumstances, but it is certainly true that much of the time and fuel that has been applied to stabilising could have been more profitably used in other directions, and it is also true that internal stress, accompanied by elastic strain and plastic deformation, a condition, for example, such as that produced by the operation of "straightening," cannot be wholly removed by stabilising or any other form of heat-treatment that is both economically and technically practicable for heat-treated nitriding steels. This rather sweeping statement has a wider application than heat-treated nitriding steels. The common practice of retempering a forging that has been straightened after hardening and tempering does not render it stress-free. The magnitude of the internal stress present may not be important for many engineering applications, but if the part is to

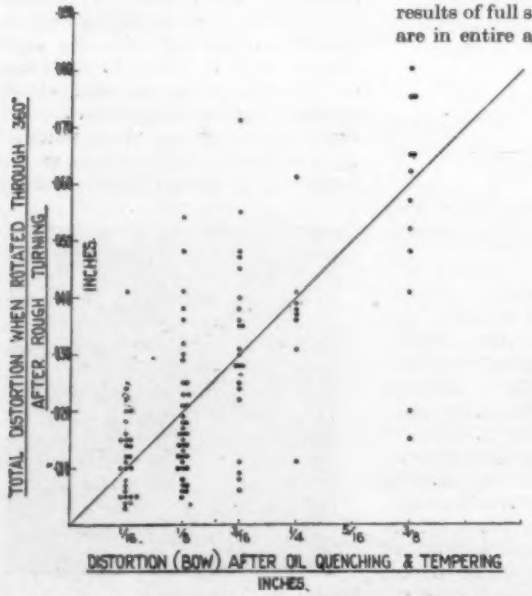
TOTAL DISTORTION WHEN ROTATED THROUGH 360° AFTER ROUGH TURNING
INCHES
DISTORTION (BOW) AFTER OIL QUENCHING & TEMPERING
INCHES

Fig. 17.—Influence of internal stress on distortion during machining.

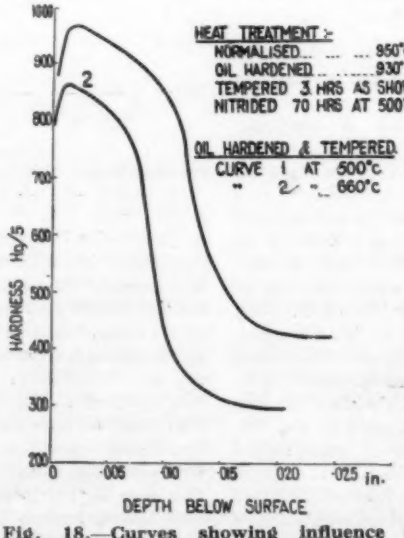

Fig. 18.—Curves showing influence of tempering temperature on maximum hardness and case depth after nitriding.

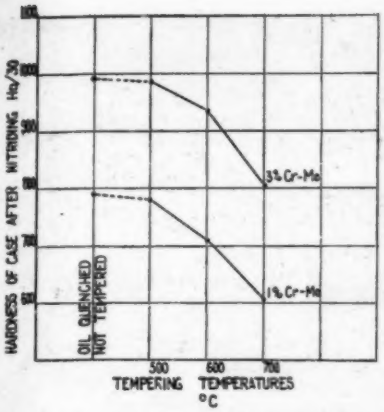

Fig. 19.—Relationship between tempering temperature and hardness of case after nitriding.

be nitrided, and assuming, as is usual, that distortion cannot, or may not be corrected by any means after nitriding the story is different; unequally distributed internal stress, if present to any appreciable extent, is then disastrous. The relationship between initial distortion and reappearance of distortion during the early stages of machining hardened and tempered crankshaft forgings that have been straightened and "stress-relieved," is shown in Fig. 17. Such crankshafts will distort again on stabilising and yet again on nitriding. It is satisfactory to be able to record that by taking suitable precautions it is possible to hold the distortion below 0·002 in. in a length of 40 in. in drop-forged crankshafts left unmachined on the webs with no more difficulty than in the far simpler examples of crankshafts forged with ample machining allowance all over.

Fig. 21.—Micro-sections of Cr-Mo-V steel nitrided at normal (left), and at a somewhat higher temperature, illustrating increase of thickness of friable layer.

Fig. 20.—Part of a plant used for nitriding Diesel engine crankshafts.

It is usually asserted that the hardness of a nitrided steel is a function of its chemical composition and that it is not affected by the previous hardening and tempering treatment, this is far from the truth. It is true that steels which rely largely on the presence of aluminium for their hardening on nitriding are not much influenced in this respect, but steels that depend chiefly on chromium as the special nitride-forming element may exhibit a considerable variation in the hardness of the nitrided layer according to the tempering temperature used. A 3% chromium-molybdenum steel, containing about 0·35% of carbon, can show a variation of almost 200 units diamond pyramid hardness number after nitriding according to the tempering temperature used; and a low chromium-molybdenum steel that has been tempered at the lowest practicable temperature, i.e., one coinciding with the nitriding temperature, may have a nitrided hardness not greatly different from a 3% chromium-molybdenum steel that has been tempered at the maximum temperature. These facts are illustrated in Figs. 18 and 19. In the presence of a nitride-forming element such as vanadium, whose carbide is only sparingly soluble in a iron at the upper critical temperature, the results may be still further influenced by the quenching temperature used. Investigations at present proceeding point to duration of tempering period as a further factor influencing nitrided hardness.

Changes the Engineer Would Like

Having given an outline of some of the advantages of nitrided crankshafts this would appear to be a suitable opportunity to voice some of its disadvantages, amongst these one might refer to three. Firstly, the steels are expensive, users would welcome a cheaper range of nitriding steels. Secondly, the treatment time is very prolonged and consequently a lot of expensive plant is required; Fig. 20 shows part of a battery of nitriding furnaces used for nitriding Diesel engine crankshafts, incidentally it was the first plant in the world to be installed specially for this purpose. Thirdly, the case produced is thin, if a steel could be developed that would "take" a deeper case without sacrificing the desirable characteristics of nitriding steels, this would be a real service to the user and one of which he would undoubtedly avail himself. It is true, of course, that the case depth can be increased by increasing the nitriding temperature (for the loss of some hardness), but this method has serious disadvantages. It increases the depth of the high-nitrogen, friable layer at the surface, as can be seen from the two specimens illustrated in Fig. 21, 2% Cr-Mo-V steel nitrided at the normal and at a slightly higher temperature. The wear and tear on the equipment is likely to be considerably greater. Taken all round it is not a procedure which commends itself.

Izod Notch Bar Value

It seems fitting to conclude a discussion, which has centred so much

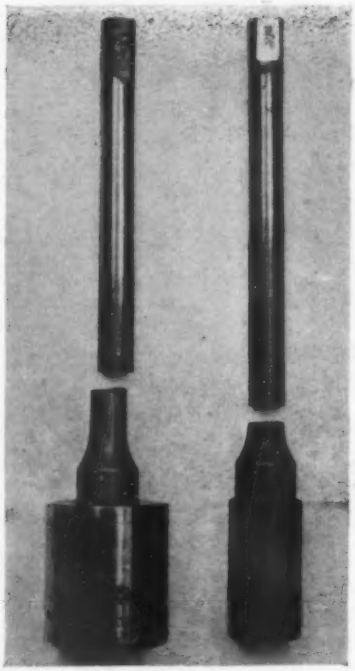

Fig. 22.—Wöhler test pieces showing influence of inclusion on position of fracture.

Fig. 23.—Fracture of the specimen shown on the left of Fig. 22. × 3½

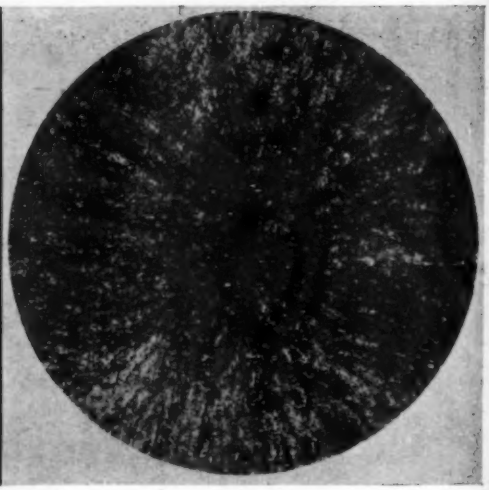

Fig. 24.—Origin of fracture shown in Fig. 23. × 50.

around crankshafts, by referring to the old debating favourites, Izod impact value and inclusions. It is often said that the Izod impact value is of no importance in a crankshaft, indeed one recalls more than one occasion on which it has been said that it was of no value at all, this, however, is going too far. Under very severe conditions of shock-loading a good Izod value is much to be desired. Experience during the war with castings used for vehicles subjected to shocks far in excess of anything a commercial vehicle is called upon to withstand left one in no doubt about the value of the Izod test as an acceptance test. To return to crankshafts: is the Izod test of value? The knowledge that a number of examples of crankshafts exist where material having an extremely low impact value behaves satisfactorily lends colour to the argument that it is not. But the fact remains that if a steel is known to be capable of sustaining a high Izod notch bar value when correctly made and treated, then the user is entitled to look for that as an earnest of the steelmakers' and forgers' care.

Inclusions

Anyone who essays to discuss inclusions in steel needs to exercise caution, and even then will possibly approach his subject with some trepidation when he recalls that users who are interested in rapid machining and good finish ask steelmakers to put inclusions in steel. This, of course, is not new, it has been the practice for more than a quarter of a century in the production of low-tensile, rapid-machining steel used for the mass production of relatively unimportant details. But recently the principle has been extended to medium and high tensile alloy constructional steels. There is, however, a difference between the presence of a large number of small, evenly dispersed inclusions, of whose presence the user is cognisant, and for which, consequently, he can make full allowance when selecting the steel for the job, and the presence of larger, unsuspected inclusions which may make themselves manifest in unexpected and inconvenient places.

Concerning the problem of non-metallic inclusions that have not been added intentionally it does sometimes seem as if this is a subject on which steelmakers speak with two voices. A student of the literature cannot help but be struck by the evident pride with which the steelmaker records his achievements in the production of clean and still cleaner steel. And yet, when one is unfortunate enough to find forgings which, on machining, show obvious evidence of inclusions the steelmaker, it seems, is not particularly worried about their possible ill-effects, the user is apt to be rather a fidgity fellow, unduly worried about trifles.

The two Wöhler test pieces shown in Fig. 22 provide an excellent example

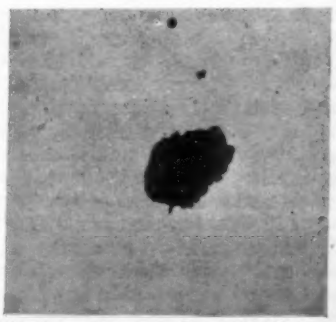

Fig. 25.—Cross section of inclusion found at the centre of Fig. 24. × 200

of what an inclusion can do. These Wöhler test pieces were nitrided specimens, both broke with the type of fracture usual for such specimens, the crack started in the core below the case, one broke in the usual place at the junction of the parallel portion and fillet, but the other broke some distance down the parallel length at a place where the calculated stress was very much less than at the position of maximum stress. Fig. 23 shows the fracture slightly magnified and Fig. 24 at a magnification of 50 diameters. The dark region at the centre of Fig. 24 suggested the presence of a non-metallic inclusion; it was disclosed by careful polishing, and is shown in cross section in Fig. 25 at a magnification of 200 diameters. The fatigue resistance was reduced 12½% by this relatively deep-seated inclusion; an excellent example of the damage an inclusion can do. It would seem that the steelmaker, in spite of his other voice, is justified in his search for clean steel.

The Importance of Ultimate Extension as an Engineering Property of Materials

By A. Fisher, F.I.M.

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One of the properties in the ductility group—percentage ultimate extension—is discussed. It is considered that a condition of specific strain is far more common in engineering structures than is generally supposed: that, under a condition of specific strain, it is quite useless to use a material with a higher modulus of elasticity, a higher proof strength, or a higher anything which moves the stress-strain curve to the left, in the hope that higher strength will result, because such a move merely results in a higher stress being generated. Under conditions of specific strain a component fractures if its available percentage extension is below that required under the imposed strain, irrespective of its strength; hence, percentage ultimate extension indicates a valuable property in any material to be used where stress concentrations exist.

THE mechanical properties of engineering materials as obtained from the ordinary tensile test are sometimes divided into two groups or classes; one group comprising those properties primarily connected with the resistance offered by the material to the imposed load, and referred to under the general heading of "strength," the other, those connected with the change of shape of the material under load and usually referred to under the general heading of "ductility."

There are some logical grounds for looking at and grouping mechanical properties in this manner; in some branches of metallurgical research, for example—particularly in the analysis of mechanical behaviour of materials when alloyed—by grouping certain properties together, and noting the general effects of different types of alloying addition, it is possible to obtain the trends of influence of the various additions, and derive valuable information as to the best direction of future research aiming at a particular result.

Not all mechanical properties can be clearly and unambiguously placed in one or other of the above two classes. For instance, proof stress is defined as the stress required to produce a definite permanent set expressed as a percentage of the gauge length. In this case it is clearly impossible to say that the measure of strength is more important than the measure of deformation, or vice versa, for both characteristics contribute to the condition of the material, by definition. It would seem that for completeness a third group at least is required—one dealing with properties connected with stiffness (*i.e.*, both elastic and plastic stiffness), and thus bringing in both strength and deformation. However, for many comparison purposes the two more or less "opposed" groups mentioned above are found sufficient.

The object of this article is to refer to the neglect and discuss the importance of one of the properties in the ductility group, namely, percentage ultimate extension, a very close relative of the more familiar property, percentage elongation. Percentage ultimate extension is, as the term indicates, the measure of total extension under tensile load. Percentage elongation, as defined by BSS.18/1938,[1] is "the percentage increase of the standard gauge length obtained by measurement of the fractured test piece." The difference between the two is obviously the value of the elastic extension, an almost negligible fraction if the elongation is of the order of 10% or 20%, but a most important and sometimes the major fraction if the elongation is under 1%.

The property of percentage elongation is determined in ordinary tensile testing simply because it is easier to measure the broken test piece after fracture than the unbroken test piece at the instant immediately before fracture, but it is the value of the ultimate extension, rather than that of elongation which is important, as will be seen in later discussion. Because there is little difference between the two values for usual elongation figures—*i.e.*, say, over 5%, and because of the ease of obtaining the percentage elongation, this property has been accepted as the most convenient and at the same time a reasonable measure of ductility. It is, nevertheless, the property of percentage extension, not percentage elongation, which the engineer really requires in strength calculations and stress analyses. He is concerned with the capacity of the unbroken material to extend under load rather than with the residual extension in the broken pieces after fracture. The distinction, as wi l be seen, is important and critical.

For many purposes, however, and in reference to the subject of the value of ductility generally, the percentage elongation and the term "elongation" can be, and often are, used without serious error in place of the more applicable and correct value and term "ultimate extension." In the following discussion, to avoid the appearance of pedantic phraseology, this common usage will be adopted where permissible, or except where accuracy is seriously violated, though the writer believes it desirable for many reasons to bring the term "extension" into greater use where dimensional increase in length under load is involved. It will therefore be in order, and helpful, to read the more strictly correct term, "ultimate extension" throughout this article, for the more colloquial and generally-used term "elongation."

Ductility and Gauge Length

In materials to be used in any engineering application where high local stresses have to be carried, where shock loads have to be borne, where notches or sharp changes in section cannot be avoided, particularly where a duplex type of structure (metallurgical or mechanical) is loaded under conditions of specific strain, good ductility is an essential requirement, if satisfactory performance in service is to be obtained.

1 British Standards Specification, 18/1938, "Tensile Testing of Metals."

Unfortunately, this is not well understood; in fact, a good deal of misconception exists in many quarters regarding the importance of ductility as an engineering property of structural materials. There are several reasons for this, one is possibly the residual influence of past experience with temper-brittle nickel-chrome steels which have good slow-speed, but negligible high-speed or impact elongation. It should not be argued, of course, that, because a good elongation is obtained on a material that fails in a certain application, a good elongation can not be an essential requirement in that application. The fact that some nickel-chrome steels may have a high elongation and yet fail by brittle fracture under shock conditions in no way proves that high elongation is not an essential requirement for good service under shock conditions.

Probably the main reason for misconception on this point, however, is the almost universal tendency to interpret mechanical properties by the medium of the standard tensile test, and to think of elongation in terms of a two-inch guage length, and nothing else. In considering or applying the principles of engineering design, there can be no greater fallacy than to think that 10% elongation means 0·2 in. extension on a 2-in. gauge length, and only that. Ten per cent. elongation in a material of duplex or multiplex microstructure may sometimes mean an extension of $\frac{1}{1000}$ in., or even less, in some particular micro-constituent; but $\frac{1}{1000}$ in. may mean failure in that constituent if it happens to be brittle, even though it may be the strongest phase present. In this case, it will have to be conceded that, contrary to the popular idea, the weakness of the chain is in its strongest link.

Gauge lengths of 2 in. or greater, are used in practical testing because of the ease with which accurate measurements of extension can be obtained, using the simple, convenient and standard extensometers. There is no reason whatever why the actual elongation obtained on a 2-in. gauge length should be considered the sole, or the best, or even a good criterion of usefulness of the property of elongation in general. In actual fact, it is in most cases the percentage extension on short guage lengths, such as 0·02 in. or even 0·002 in., which is critically important in the design and performance of engineering structures or structural parts, and the consequences of ignoring or not realising this fact may be very serious.

It is erroneous and misleading to suggest that because a structural element will not be stressed to the point where it extends 0·2 in. (10% in 2 in. gauge length) a figure of, say, 10% elongation as a specification figure is excessive, yet how often are such suggestions made in specification or material and design discussions. "Percentage elongation," it is protested, is the elongation corresponding to the breaking stress, which is not reached until after the ultimate stress is exceeded, and it is not intended to stress the structure to the breaking stress: it is further pointed out that in fact the structure has been, or will be, designed to a figure something less than the proof stress or to some other safe working stress which is only a fraction of the ultimate, and therefore an elongation equal to the "percentage elongation," could not possibly be produced in any component part of the structure by the loads to be applied in practice.

This seems plausible enough on the face of it, but it is a line of reasoning which can be very easily misapplied, and so lead to wrong conclusions. It may be quite correct that if the maximum stress developed in any component of a given structure is limited to the 0·1% proof stress, then it can hardly be said to matter, for the safety of that structure, whether the elongation of the material at the ultimate or breaking stress is 1·0% or 10·0%,[2] but if this statement is used to argue that ductility generally, and ultimate extension in particular, are unimportant factors in design considerations, it is being misapplied. Certainly, at the present time, in design little use is being made of any data which may be available on the ultimate extension values of different materials, but this is because of the lack of understanding of the importance of such data, and the lack of knowledge of how to use it. Even in straightforward and standard cases such as simple tensile, bending or torsion members of uniform cross-section, the engineer does not know what the true stresses really are. Admittedly, certain figures can be obtained by the use of certain simplified formulae, such as $M/I = f/y = E/R$ for beams, but these give merely approximate solutions. In the derivation of the beam formula just quoted, for example, no fewer than 10 basic assumptions are involved.[3] In any member in which a change of cross-section occurs, the standard formula becomes inapplicable, and the error introduced by their use in such cases may amount to several hundreds per cent.

The point is therefore, that because of the serious inadequacy of present design formulae, it is impossible to know whether the maximum stress developed in any component of a given structure is actually limited to the normal 0·1% proof stress at all points, or, in components with changing cross-section, whether that stress is being exceeded by several hundreds per cent. at some point.[4] But it is exactly in these latter cases that the percentage ultimate extension is of the very greatest importance. In fact, so far from ultimate extension being an unimportant property in design considerations, it may be, under a condition of specific strain, the low ultimate extension value of a portion of the structure which is itself limiting the stress and the load that can be applied, in which case a higher ductility would have enabled a more efficient design to be used. This point is discussed later. Clearly, the value of elongation as an engineering property of materials, is a matter that urgently needs to be re-assessed.

The attitude of the "down with elongation" school to the question is well expressed by Brearley,[5] as follows: "After using a tensile machine for a few years, the routine tester becomes incapable of thinking about its limitations and misrepresentations. He is right in regarding it as one way, as good as another, of measuring the hardness of steel, but he cannot be weaned from the idea that elongation per cent. means something, and he cannot be convinced if his mind is not open to conviction. The value of testing is to assess the likely resistance of material to destruction. The elongation per cent. is obviously a factor in the destruction of the smoothly-machined test piece which is being pulled apart in the tensile machine. But not one piece in a thousand, bearing the stress of engineering service, faces destruction under tensile-testing machine conditions. When the remnants of a blitzed factory or the debris of a

2 "How Important is Elongation in Aircraft Design?" A. Fisher. *Magnesium Review and Abstracts*. Vol. III. No. 3. April, 1945.
3 "Beam Design in the Curvilinear Range of Stress." A. Fisher. *Metallurgia*. March, 1945.
4 "A Discussion on the Notch Impact Test." A. Fisher. *Metallurgia*. September-October, 1945.
5 "Testing Steel." Harry Brearley. *The Engineer*. June 9 and 16, 1944.

collision or the jumble of a collapsed building are looked over, there are many steel parts smashed, torn, twisted and bent, but all gave way before their broken parts had stretched to the average of the tensile test piece. The percentage elongation is recorded because it can be easily observed; it is measured, and, being measured, it is supposed to be significant, though no one says what it signifies in the sense that it was significant for the maker of wrought iron.

The engineer desires hardness combined with toughness; a sufficient amount of the one, and as much as he can get of the other. By toughness, he means ability in material to bend and distort and tear, rather than break short, however unfavourably it might be shaped or stressed. He accepted the ductility of a smooth cylinder, stressed slowly, as a measure of toughness. That was a mistake."

Thus, not only do energetic salesmen of high-strength alloys, designers with a passion for pushing up the proof stress, and unimaginative test-house managers mechanically producing millions of routine test figures, but also some metallurgists, help towards the maintenance of the present completely wrong idea that while a good elongation is something about which a wire-drawer might be expected to rejoice, it is not really anything about which an engineer-designer need concern himself. This article is intended to provide evidence to the contrary.

Percentage Extension under Tessellated Stress Conditions

In any piece of solid metallic material subjected to an increasing tensile load, an increasing deformation will take place, at first elastic in nature, afterwards plastic. A stress-strain curve plotted from a tensile test of the material will provide the relationship between the load and the deformation. If the material is substantially "homogeneous" throughout, the test bar being specially made, say, from a single crystal, then one stress-strain curve may be taken as approximately representative of the load-deformation relationship for any part of the test piece. There may be a very slight difference in the behaviour of a portion of the test bar near the radius at the shoulder and that of a portion near the centre of the gauge length, but this is unimportant for the point being discussed. There may also be differences on an atomic scale at positions where slip is taking place by the relative movements of atoms, but insufficient is known about the spheres of influence and the magnitudes of stresses involved on the atomic scale to be able to say anything relatively useful, quantitatively.

If the piece of material being tested consists not of a single crystal, but of a coherent mass of crystals—*i.e.*, is polycrystalline in character—the usual conditions for structural materials—then not only will one stress-strain curve not be sufficiently representative of the load-deformation relationship for all parts of the piece, but the stress-strain curve obtained in a tensile test of the polycrystalline piece will not correspond with that obtained from the single crystal of the same basic material. The reason for this is partly the local stiffening influence of the material in the neighbourhood of the grain boundaries—which is structurally unfavourable for normal homogeneous slip—and partly the varying orientation from crystal to crystal, diagrammatically represented in Fig. 1, which likewise interferes with general homogeneous slip.

No material is perfectly isotropic, and most metals, even in single crystal form, are appreciably anisotropic, the properties varying considerably in different directions. Therefore, in the direction corresponding to that of the applied stress—that is, axially along the test bar—the deformation of the separate crystals in the polycrystalline piece will tend to vary, and the result will be that for any given overall strain on the test bar, internal local stresses of different magnitude will be generated in different crystals.

Fig. 1.—Varying crystal orientation.

These stresses have been named "tessellated" stress by Lazlo,[6 7 8 9] who has investigated their occurrence and the effects, on such stresses, of such factors as thermal changes and phase transformations. He shows, for example, that if a polycrystalline specimen of pure iron with crystals in certain relative orientations is subjected to a homogeneous tension below the elastic limit, tessellated stresses equal to + 35% and − 37% of the original primary tensile stress are developed within the specimen. He also points out that in the thermal effect in pure zinc, a change in temperature may cause the development of tessellated stresses equal to 274·5 lb./sq. in. in tension, and 173·5 lb./sq. in. in compression per degree centigrade, due to crystal anisotropy alone, independent of any other imposed stress system.

It is clear that if a tensile specimen of anisotropic material consists of an aggregate of crystals in different relative orientation, tessellated stresses will develop in the specimen on loading, such stresses—as pointed out by Lazlo—being related to the elastic constants, if the elastic limit for the material is not exceeded.

It is possible, however, in materials of duplex microstructure (where the different structural constituents have usually appreciably different stress-strain curves) for the tessellated stresses to pass far beyond the elastic limit for one constituent, while the homogeneous applied stress—*i.e.*, the theoretically-distributed stress, remains nominally still well below this point. It is even possible to reach the breaking stress in one of the constituents, while in the other the stress has not exceeded a safe working figure: everything depends upon the relative shapes of the stress-strain curves for the different constituents.

It is for such reasons as these that the standard text-book type formulae are unable to tell the engineer what the true stresses really are, in the simple tensile or bending member of uniform cross-section. Still less do they assist him in cases where the cross-section varies.

The Condition of Specific Strain

Fig. 2 represents in diagrammatic form the duplex microstructure of a specimen to which a tensile load is being applied. It is quite typical of a great many alloy structures in which a comparatively brittle "beta," or second phase, appears in "pools" and "networks" in the grain boundaries of a ductile "alpha" phase. In the diagram, the beta phase "B" is seen to be "locked"

6 "Tessellated Stresses." Part I. F. Lazlo. *J.I.S.I.* No. 1, 1943.
7 "Tessellated Stresses." Part II. F. Lazlo. *J.I.S.I.* No. II, 1943.
8 "Tessellated Stresses." Part III. F. Lazlo. *J.I.S.I.* No. II. 1944.
9 "Tessellated Stresses." Part IV. F. Lazlo. *J.I.S.I.*. (Advance copy. November, 1945).

or " keyed " in the alpha phase " A," so that when the latter extends, due to an application of load, the beta phase is compelled to extend also, to a like amount. This is a condition of " specific strain " on the beta. The composite or resultant extension on the specimen will, of course, not be quite the same as if the structure of the material had been completely alpha, as the beta will exert a stiffening effect on the alpha to an amount depending on relative quantities, the moduli of elasticity, and the effective plastic stiffness. The alpha phase, however, being much more ductile than the beta, can deform to an extent quite beyond the capacity of the latter, and the result of a continued extension of the test piece is comparatively early fracture in the more brittle beta, causing a sudden increase in stress in the alpha, and an acceleration of final failure.

If the tensile stress-strain curves for two such constituents are known, and are plotted on a common diagram, the instantaneous state of stress in both constituents can easily be determined, to a close degree of approximation for any given common strain. Fig. 3 shows the two stress-strain curves for the constituents " A " and " B " plotted from the same origin and on the same scale. For a given composite strain on the test bar, say, " e," there is a corresponding stress " a " in the alpha where the vertical ordinate through " e," cuts the stress-strain curve. As, through the interlocking of the structures, the strain in the beta must also be equal to the value " e," whatever stress may be produced (condition of specific strain), the same ordinate from " e " produced to cut the beta stress-strain curve at " b " will enable the stress figure for the beta phase to be obtained for that strain. The figures corresponding to " a " and " b " are therefore the instantaneous stresses in the two constituents A and B for the common strain " e." The excess of " b " over " a " will be noted. It will be seen that the application of load and the consequent strain in the alpha causes the stress to rise rapidly in the beta, and failure will occur first in this constituent notwithstanding the fact that, as shown by the stress-strain curves, the beta phase at the point of failure is considerably stronger than the alpha, so far as stress capacity is concerned. In this way does the weakness of the metallurgical chain show itself to be in the strongest link.

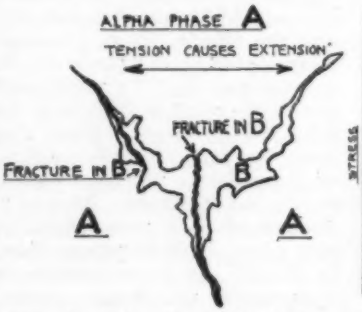


Fig. 2.—Beta phase fracturing under specific strain.

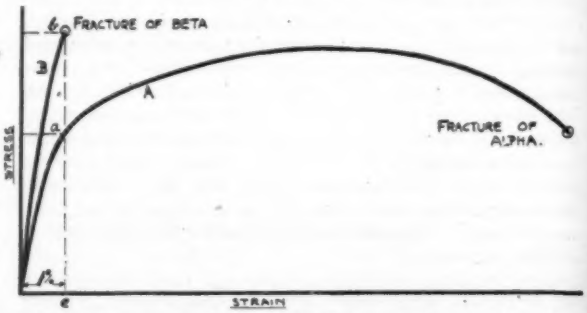


Fig. 3.—Stress-strain curves of alpha and beta in material of duplex structure.

It appears to be clear, therefore, that in any material having a duplex and interlocking type of structure in which the constituents possess different elastic moduli and different stress-strain curves, a condition of loading resulting in a certain strain in the major constituent (alpha) causes the minor constituent (beta) to assume a corresponding strain (specific strain type of loading) which may be entirely beyond its capacity, and the minor constituent fractures if its available extension is below that required under the imposed specific strain, *even if its ultimate tensile or its breaking stress is higher than that of the major constituent.*

From one point of view it would seem possible to say that the material fails because the breaking strength of the beta is less than would be required to withstand the strain imposed, but it should be pointed out that under the condition of specific strain, any increase in strength, no matter how great, would be too little if the available extension were insufficient. Therefore, under this condition, a much more common one than is usually supposed, it is the percentage extension of the material, and not its strength which is the criterion of failure.

Percentage elongation values have previously been used by engineers largely " for information only," and there seem to be occasional tendencies, as already mentioned, to ignore them altogether. The foregoing considerations would appear to be sufficient to provide reasonable grounds for treating " ultimate extension " with more respect. Admittedly, the above example refers in particular only to the separate constituents of a duplex micro-structure, but it should be evident that whether the separate parts are constituents in a micro-structure or components of an engineering structure, so long as there are the two factors (*a*) a specific strain existing in the structure, (*b*) different available percentage extensions in the separate constituents or components of that structure, then fracture in the less ductile element on account of lack of elongation rather than lack of strength, is at least a possible mode of failure.

Obviously, both the engineer and the metallurgist are concerned. The latter makes a less number of calculations for strength than the former, because his structures are less definite in shape and are not " designed " by calculation to withstand given loads, but he should at least have the right kind of concept, and use correct bases of comparison, in order that he can exercise proper control, and guide his alloy development and construction in the right direction.

The position can be put quite simply for both engineers and metallurgists. If loading is of the " specific stress " type, it is sufficient to design (compare) on a basis of stress, using whatever factors of safety are considered to be suitable, in the usual way. If loading is of the " specific strain " type, or a condition of specific strain is generated, it is not sufficient to design (compare) on the usual stress basis, but the percentage extension of the material concerned must be compared with the actual local extension produced in the structure, on the appropriate gauge length, or alternatively, the actual stress at the " specific strain " concerned may be

obtained from the complete stress-strain diagram for the material, and checked back against the design stresses and safety factors. Therefore, the first thing to be done by either the engineer or the metallurgist in any consideration or calculation of strength, before even the basis of design can be determined, is to examine the loading conditions, and see whether the structure, or any part of it, comes under conditions of specific strain. To analyse the loading conditions within the material the metallurgist must examine the microstructure, and determine the properties of the separate phases.

Examination of microstructures of pulled tensile test bars of material with duplex structure will almost always show extreme failure conditions in the brittle constituent. If test bars of such material are strained to different amounts below the fracture point, it is easy by micro-examination to determine the stress at which fracture in the beta takes place, and so obtain information which can be used to guide compositional changes, heat-treatment procedure, or other factors which may possibly influence the amount, form or properties of the beta phase. Fig. 4 shows an example of fracture in the beta constituent occurring before failure in the alpha in a magnesium alloy tensile test piece.

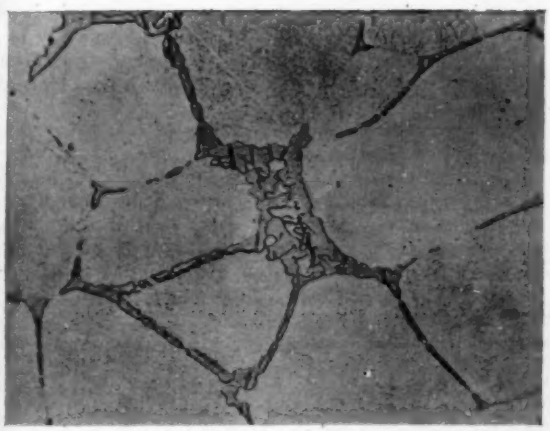


Fig. 4.—Slow fracture in beta phase, under tension.

The importance of knowing the properties of the separate structural constituents in any systematic investigation aiming at the improvement of existing alloys, or the discovery of new alloys with high mechanical properties, will be apparent. It is in fact impossible to carry out such new-alloy research on a truly rational basis, unless the properties of the separate phases are investigated. Very few, if any, principles of alloys research appear to have been laid down up to the present, but this might perhaps be reasonably accepted as being one of the most essential. Knowing the properties of the separate phases, those ranges of composition producing, or likely to produce appreciable quantities of brittle constituents can then be avoided, or, alternatively, appropriate physical, mechanical or heat-treatments may be applied to modify the quantities or properties of the phase responsible.

This difference in criterion of failure under conditions of specific strain, as compared with specific stress, is of the highest importance in the design of structures; it is, in fact, fundamental, and as mentioned previously, it should receive first consideration from the designer, in order that he may determine the type of design formulae to use. In the past, altogether too little attention has been paid to this preliminary but essential aspect of design.

Tensile Stress Distribution and the Necking Phenomenon

When an ordinary test piece of uniform cross-section between the gauge length marks is subjected to tensile load, the tensile stress on the cross-sectional area, when considered in a longitudinal direction, is uniform also (except for the influence of tessellated stresses) until a considerable proportion of the possible plastic strain has been produced. Up to the point of commencement of necking—*i.e.*, while the reduction in area is appreciably constant along the gauge length, any local slip tending to cause local reduction in area is accompanied by local strain-hardening, which tends to prevent further slip at the same point. So long as the rate of strain-hardening at any point is greater than the rate of reduction in area, the total resistance at that point after slip will be greater than before, and further slip will then tend to occur elsewhere in the piece where similar local strain-hardening has not already taken place. In this way under increasing load the material hardens up reasonably uniformly over the whole volume, and the reduction in area remains constant along the gauge length, the resistance to slip gradually increasing over the whole test piece, and the stress-strain curve following an upward path.

In metallic materials the rate of strain hardening gradually falls as slip proceeds—except in special circumstances, where other types of deformation co-exist—and the stress-strain line, while following an upward path, curves over away from the straight line representing the Hooke's-law relationship of elasticity, to produce the characteristic stress-strain curve for the material. (A great deal of ink has been expended recently in discussing [10] [11] [12] [13] the point as to whether there is a real family resemblance between the stress-strain curves for different materials—*i.e.*, whether there is some general fundamental law governing the fall in the rate of strain hardening. Up to the present, blood relationship has not been proved).

When the rate of strain hardening falls to less than the rate of reduction in area, then, obviously, the total resistance at the point where slip is taking place will begin to decrease, with the result that further slip and accompanying reduction in area will continue at the same location, instead of being forced to occur elsewhere. Consequently, the stress-strain curve ceases to follow an upward path, and begins to curve downwards from the point where the two rates coincide: this is the point of maximum load, corresponding to the stress defined by BSS 18/1938, as the "ultimate tensile stress." After this point has been reached, the more that local slip takes place the lower the local rate of strain-hardening falls below the local rate of reduction in area, and the lower the total resistance becomes. The lower

10 "A Renaissance of Mechanical Properties." A. C. Vivian. *J.I.M.E.* September, 1944.

11 "Significance of Tensile and Other Properties." H. O'Neill. *J.I.M.E.* September, 1944.

12 "The Shape of a Material's Reaction to Force." A. C. Vivian. *Metallurgia.* March, 1945.

13 "Strength in Technical Parlance." A. C. Vivian. *Metallurgia.* 1944.

the total resistance, the greater tendency there is for still further slip to occur at the same point, and so necking commences and proceeds until the piece fractures. Once necking has started at any point, of course, no further strain-hardening occurs at those portions away from the neck.

The cause of necking appears, therefore, as a mere coincidence between the (falling) rate of strain-hardening and the rate of reduction in area occurring as the test piece is extended, and is thus established as an accidental, not a fundamental, phenomenon.

Percentage Extension under Stress Concentration Conditions

It will be clear, then, that on a standard test piece, up to the point of commencement of necking, and ignoring tessellated stresses for the time being, the tensile stress on the cross-section, considered along a longitudinal direction, will be reasonably uniform, and correspond to "load per unit area." This is not so, however, on test pieces containing a notch or sudden section change: in these cases, while the loads—and nominal stresses—are still very low, and apparently well within the elastic limit, variation may occur in the actual tensile stress at different positions along the test piece, not only beyond that represented by the change in load per unit area, but also beyond the elastic limit, or even the ultimate tensile stress.

It is here that the percentage extension figure becomes very important. In general terms, the condition may be described as follows: In a strained notched test piece, the overall strain under any given load is largely determined by the main dimensions, but the notch may prevent this overall strain from being uniformly distri-

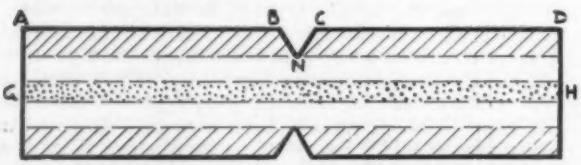


Fig. 5.—Showing portions of notched test bar which extend different amounts, and so generate stress concentration at base "N."

buted longitudinally, in some cases intensely localising it, and thus producing a state of specific strain, with the result that the required percentage extension at the notch, on the effective gauge length there, is very greatly increased. In this case, in contrast with the component of uniform cross-section, it will be seen that it is unnecessary to have two different materials, or material with different stress-strain curves, in order to produce a state of high local specific strain. The condition can be obtained in a piece of homogeneous material by a system of internal differential strain, such as is created or induced by a notch or hole.

Fig. 5 shows a tensile test piece with a deep Vee notch with small root radius. When gripped at the end and subjected to a tensile load, the piece will extend as a whole. The portions AB and CD (shown cross hatched) will, however, owing to the intervening notch space BC, clearly not be subjected to such a degree of tension as the paraxial portion GH (shown dotted), and will consequently in themselves not extend the full amount. but will tend to move bodily end-wards when the piece extends as a whole. As the total extension between A and D is maintained approximately the same as that between G and H, the lack of extension in the separate parts AB and CD causes an essential increase in extension on the portion between them, at the base of the notch N, in order to produce the required total extension between A and D, equal to that between G and H.

Here, then, we have a condition where the local extension in a piece of material is produced or controlled by the extensions occurring in some other part of the piece—that is, a condition of specific strain—and again we have, in consequence, extremely high stresses generated at the site of local extension, and failure if the ultimate extension is less than the required extension, irrespective of the local or general strength of the material.

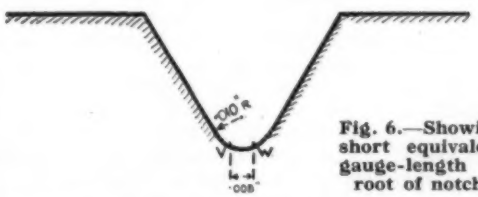


Fig. 6.—Showing short equivalent gauge-length at root of notch.

It is not difficult to see that the increase in extension at the base of the notch, though possibly small in itself, as an absolute amount, may be equivalent locally to a very high percentage extension, while at the same time the overall percentage extension of the whole test bar is quite low.

For example, in the Vee notch of 0·01 in. radius, shown in Fig. 6 on an enlarged scale, the greater part of the extension will take place on the portion VW, which is only 0·008 in. in length, and a local increase of only 0·001 in. on this gauge length is equivalent to an extension of 12%. If the local extension at the base of the notch under a given load were as little as 0·00006 in. (1·0% of the effective gauge length), while a material like that represented by curve A in Fig. 3 would survive, one like that represented by curve B would fail, because the ultimate and breaking strength of the latter, although the stronger material, is reached at 1·0% extension. A crack is compelled to form when the ultimate extension of the material has been fully expended locally, however small the gauge length may be.

Thus, with notched components, as with material of duplex micro-structure, it is the ultimate extension of the material rather than the strength, which determines the point of initial failure.

The frequently-occurring paradox of high strength material with low ultimate extension—or elongation—failing at lower loads than lower strength material with higher ultimate extension[14 15 16 17]—or elongation—needs no further explanation. Wherever local stress is dependent on, or governed by, such factors as change in cross-section, change in structure or change in properties at discontinuities or grain boundaries, then the local stress must increase as the resistance of the material to deformation increases, because of the existence of specific strain conditions.

14 "Brittle Fracture of Metals." B. P. Haigh. *Engineering*. Vol. 130. 1930.

15 "Prevention of Failure of Metals under Repeated Stress." Batelle Memorial Institute. 1941.

16 "Influence of Chemically and Mechanically Formed Notches." MacAdam and Clyne Research Paper R. P. 725. *Bureau of Standards*. Vol. 13. 1934.

17 "Work Hardening and Notch Sensitivity of High Strength Steels." Pevaner. *J.I.S.I.* Translation, 1945.

Materials of high strength have almost invariably a lower elongation value than the same material in a condition of lower strength, consequently, a loss rather than a gain in strength may be expected to result from the substitution of the former for the latter in any application where specific strain conditions exist. Conversely, a gain in strength can often result from the use of lower strength, but more ductile material. It is well known, both from extensive laboratory tests and—often disastrous—results in service, that high-strength materials have higher notch sensitivity than the same materials in a condition of lower strength. The property of ultimate extension is one of the most critical factors in notch sensitivity phenomena.

Percentage Extension Under Impact Conditions

The ability of a component to deal with loads of an impact type depends on the capacity to absorb the kinetic energy of the impacting body during the short time the load is applied. In producing the necessary resistance to this type of load, both force and distance—moved—through become factors of the product, and therefore the available extension of the component is just as important as the allowable stress. The problem is thus in reality one of resilience, and can be considered under two headings, (*a*) normal loads, (*b*) abnormal loads.

(*a*). If the deflection under load is to be limited to a given safe figure, which will generally be the intention for normal repeated loads to be sustained without damage this deflection becomes "available extension," and the ultimate extension of the material—corresponding to the breaking strength—enters into the consideration of strength only so far as a condition of specific strain may be generated locally in the component, as explained in previous sections of this article. Under the given condition of limited deflection for normal loads, and assuming homogeneity and absence of stress concentration, theoretically the ultimate extension does not enter in as a factor dependent on, or by virtue of the nature of the load, and if the loading conditions, stress-strain curves and structural conditions are fully known, the problem should be capable of being dealt with as one of specific stress.

In practice these assumptions are not realised, and the possibility of a condition of specific strain being developed is considerable, as has already been emphasised in connection with static loads, so that actually, percentage extension remains a factor of great importance. Where stress concentration exists, that is, in any part with changing cross-section, impact loads are correspondingly more dangerous, because then there is possibility of combination between the notch effect and the velocity effect, the latter acting as an intensifier of the former.

(*b*). Serious impact loads are generally of abnormal rather than of normal occurrence, however, and the question here arises as to whether it is desirable from some point of view to use a material with sufficient elongation to hold together after absorbing the full impact energy—possibly deforming excessively, but not breaking—or whether it is better to take advantage of whatever increase in proof strength may be obtained by a sacrifice in elongation, and have the structure fall to pieces on receipt of an abnormal impact load. In some materials the two properties are related in a balanced manner, so that as one goes up the other comes down, and the highest values in both cannot be obtained together.

It has always seemed to the writer desirable to have a key component hold together, even with excessive deformation, if this will save the remainder of the structure from severe damage, and prevent loss of life: in most cases probably the deformed component will be fairly easily replaceable, and in an aircraft, whether the abnormal impact occurs as a result of shell-fire, forced landing or other similar contingency, it is probably better that the whole structure should hold together over the critical period, than that it should disintegrate immediately. Under abnormal loads arising in such manner as mentioned above, the limitation of deflection under load necessarily ceases to exist, and the percentage extension becomes the "available extension." To use a material with sufficient extension to provide the necessary energy-absorbing capacity, may or may not entail some slight increase in weight on account of the correspondingly lower proof stress which may be expected (unless new lightweight materials can be developed with both high proof and high elongation): in high-altitude fighting aircraft even a small increase in weight may be prohibitive; but in civil aircraft it may be considered worth while for the extra safety gained.

So far as energy absorbing capacity is concerned, the advantage which can be gained by using material with high elongation, rather than high proof stress may be very considerable,[2] because there are much greater differences in percentage elongation than in proof stress or fatigue strength figures in materials likely to be used on any given aircraft component.

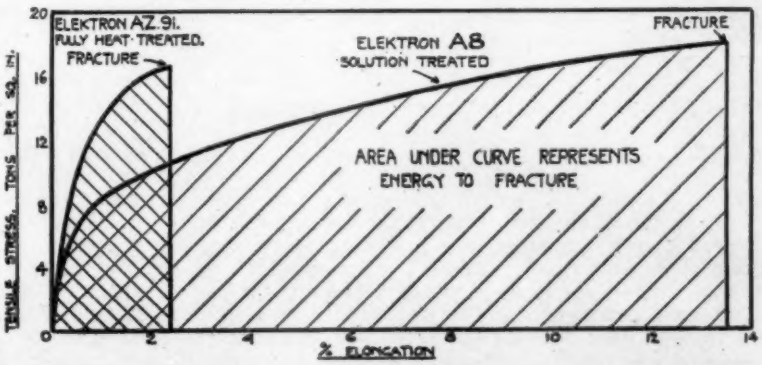


Fig. 7.—Comparison of energy absorbed in fracturing Elektron alloys.

As an example, Fig. 7 shows a comparison of the total energy-absorbing capacity—given by the area under the stress-strain curve—for the ultra-light alloys Elektron AZ 91, in the condition as fully heat-treated (D.T.D. 281), and Elektron A8, solution-treated (D.T.D. 285). It will be seen that the energy-absorbing capacity of the latter (measured at the speed of tensile testing) is many times that of the former. At high speeds of application of load, the ratio is not very greatly altered.

The disadvantage—if any—in weight which follows the use of high-elongation materials depends on the basis of design. For example, if the proof stress basis is used for components to be made in one of the above two Elektron alloys, the specification 0·1% proof stresses of which are—D.T.D. 281, 6·5 T.S.I., and D.T.D. 285, 4·5 T.S.I., then the weight disadvantage will be roughly in these proportions. If the fatigue strength basis is used, then there is no disadvantage in using the high-elongation material but a 10% gain, the notched fatigue strength figures being 4·5 and 4·0 T.S.I. for the Elektron A8(S.T.) and Elektron AZ91(F.H.T.) respectively. In this case, the high-elongation material is obviously the one to use, as a simultaneous gain in both strength and lightness may be obtained.

In Conclusion

The burden of the foregoing remarks is that a condition of specific strain is far more common in engineering structures than is generally supposed: that under a condition of specific strain it is quite useless to put in a material with a higher modulus of elasticity, a higher proof strength or a higher anything which moves the stress-strain curve to the left, in the hope that higher strength will result, because such a move merely results in a higher stress being generated: that under conditions of specific strain a component fractures if its available percentage extension is below that required under the imposed strain, irrespective of its strength: that percentage ultimate extension is a valuable property in any material to be used where stress concentrations exist: that higher percentage extension may mean greater safety in the event of abnormal loads, and not necessarily greater weight.

It is hoped that these notes may provide some food for thought and consideration amongst what may perhaps be termed the low-elongation school of engineering design, and possibly succeed in stemming to some extent, at an early stage, any movement which might be made towards the wholesale reduction of elongation figures in order that higher proof stress figures may be reached.

In the writer's opinion, the inauguration and adoption of a fashion of low-elongation high proof-strength materials in aircraft construction would be likely to lead to a series of costly failures. There are other opinions, of course, most of which seem to be diametrically opposed to the writer's.

Urals in The New Five Year Plan

By I. Andronov

SO far over 12,000 mineral deposits, which include almost all chemical elements, have been found in the Urals. These immense reserves of raw materials ensure the development of large-scale production in the Urals, especially since the concentration of raw materials is accompanied by rich power resources: coal, oil, etc.

Although this wealth has been tapped for over two centuries, and many metallurgical and other enterprises have been built, yet during and after the first world war the Urals industry declined. In 1920 as compared with 1913 industrial production had diminished to less than a sixth; the production of pig iron had declined by 11%, coal went down by 23%, and the generation of electricity by 85%.

The situation became entirely different during the second world war. Then the Urals was rapidly developed and embraced the major industrial districts of the Soviet Union. In 1945 industrial output more than trebled as compared with 1940; the output of steel almost doubled as did the output of coal and the production of electricity.

The Five-Year Plan for the restoration and development of Soviet national economy for 1946–1950 provides for a still greater development of the Urals industry. As compared with 1940, pig-iron smelting is to be increased two and a half times, coal-mining by 2·7 times and the output of oil by 3·2 times. The power base in the Urals is to increase by 1,300,000 kilowatt and a big hydro-electric power station is being built on the Kama River.

Construction is now in full swing in the Urals. New factories are being built, new open-cast coal cuts constructed and new oil wells drilled. Mines are being sunk at iron, non-ferrous and various rare metal deposits. New blast and open-hearth furnaces, rolling mills and coke batteries are under construction at Magnitogorsk, Chelyabinsk, Tagil, Zlatoust and Chusovaya. The largest metallurgical works in the Soviet Union is being built in the southern Urals at the rich ore deposits in Orsk, Khalilov district. A special plant is under construction for the production of electric steel. The productive capacity of Bogoslovsk, Bakal, Blagodat and Vysokayagora iron mines is being doubled.

The Urals Engineering Works which during the war was producing munitions for the Army is now being reconverted to turn out locomotives; machines for steel plants, coal and oil industries, machine-tools, instruments and other equipment for new power stations. Tank factories have now taken up the producton of transport machinery, tractors, heavy drilling machines and similar equipment. Chemists are also extending production.

The Bogoslovsk and Volochansk collieries, the Polunochny manganese mines, a large metallurgical centre arising on the base of the local coking, coal and iron-ore of Ivdel deposits, all of these are new industrial districts of the northern Urals. Here are immense reserves of coal, iron, bauxite, manganese, copper and gold.

In Ivdel, the most northerly district in the Urals, the people are still travelling on reindeer and dog-sleds and in the winter hunters make their way into the forest thickets; Mansis, the northern people, hunt in these places and thousands of sable, marten and squirrel skins are shipped from there to all parts of the world. And now geologists and other scientists and builders have come to the land of the Mansis. A railway has been built in the taiga. Iron and gold mines are working. On the map of the northern Urals appear new towns and hamlets: Ivdel, Novobe, Iezovsk, Severourals, Veselovsk, Volochansk. A permanent base of the Academy of Sciences, the Urals Branch, has been established in Ivdel and scientists are making a thorough study of the northern extremity of Kamenny Poyas, a land of precious stones and other minerals, another Urals in miniature.

On the whole of the Kamenny Poyas stretching for 932 miles from its lower foothills to Trans-Arctic Ivdel a new industrial Urals is being built which will be one of the main industrial districts of the Soviet Union, and of great importance for its entire national economy.

Control of Shrinkage Porosity and Mechanical Properties in Chill Cast Phosphor Bronzes, Leaded Bronzes and Gun Metals—I

By W. T. Pell-Walpole, B.Sc., Ph.D.
Gibbins Research Fellow, Metallurgy Department, University of Birmingham

The variables in the chill-casting process which affect the extent and distribution of shrinkage porosity have been examined for phosphorus bronzes, gun metals and leaded bronzes. The rate of pouring is the most important factor, and freedom from the most harmful types of shrinkage porosity is obtained when the rate is the lowest which will fill the mould without cold shuts. This critical rate is related to the shape and size of the ingot by the formula

$$R = KP$$

where R is the pouring rate, P is the periphery of the ingot cross-section (or in the case of cored sticks, the sum of the peripheries of mould and core), and K is a constant, which for constant mould conditions and pouring temperature, depends on the alloy composition. Values of K were determined for all standard bronze compositions.

Mechanical properties of these bronzes in general vary with casting conditions in the same manner as density—i.e., optimum values are obtained, uniformly across the section, when the pouring rate is the minimum which will fill the mould without cold shuts. Slight variations in properties are also produced by the effect of pouring and mould temperature on the distribution of the brittle compound phases.

Introduction

THE soundness and mechanical properties of bronze ingots are determined primarily by the extent and distribution of shrinkage cavities, provided that the metal has been freed from harmful gases during melting, and cast under such conditions that no gas is absorbed during pouring or solidification. It has been shown previously[1] that in straight tin bronzes the fine intercrystalline fissure is the most harmful type of cavity. The extent of this form of porosity is determined by the length of the pasty zone in the ingot during solidification, in relation to the ability of the residual liquid to feed freely through the interstices between the growing solid crystals, so as to compensate for the contraction on solidification.

The length of the pasty zone is controlled by the relation between the freezing range of the alloy and the temperature gradients set up in the ingot during solidification. The former is fixed by composition (except for minor variations with the rate of cooling), but the temperature gradients are determined by a number of variable factors in the casting process which can be brought under control. These are chiefly, the method, rate, and temperature of pouring in relation to the size and shape of the ingot, and the mass and thermal characteristics of the mould.

The extent to which the contraction on solidification is compensated for by internal feeding depends on the direction in which solidification occurs, and on the characteristics of the alloy. The direction of solidification is of prime importance, since this determines to what extent a continuous supply of "feed liquid" will be available. The ideal condition is unidirectional solidification of the ingot from the base upwards—*i.e.*, the liquidus and solidus isotherms remain horizontal. The other extreme, giving the most unsuitable conditions, occurs when freezing proceeds inwards from all surfaces of the ingot. Under these conditions the liquid metal soon becomes entirely enclosed by a solid crust, so that no mass contraction can occur and the feeding from the centre of ingot outwards results in very severe porosity in the interior of the ingot. The total volume of the cavities so formed will be equal to the volume contraction of the whole ingot over the temperature range from liquidus to solidus.

In normal casting practice, the conditions of solidification of chill ingots are generally intermediate between the two extreme cases given above—*i.e.*, solidification commences from the bottom and from the sides of the ingot. With these conditions continuous feeding of the contraction in the outer layers of the ingot is possible, and additional feed liquid is generally added to the top of the ingot to compensate for this and to prevent freezing commencing from the upper surface. Towards the end of solidification, however, "bridging" occurs by the joining of crystal surfaces growing from opposite sides of the mould, thus impeding the feeding of the central portions of the ingot. The extent to which this occurs will depend on the mass and shape of the ingot, on the method and conditions of pouring, on the mould characteristics, and on the constitution and crystal habit of the alloy.

The ease with which the molten alloy can feed through the narrow interstices between the growing crystals will be determined by the physical characteristics of the alloy, by the relative amounts of liquid and solid present at the lowest temperature of solidification, and by the amount and nature of solid inclusions such as oxides, sulphides and slag particles, present in the liquid. In the case of the various bronzes, the general type of constitution and crystal habit is common to all compositions, but the relative proportions of liquid left at the lowest temperature of solidification varies very considerably as between the straight low-tin bronzes at one extreme and the highly-leaded bronzes at the other. The type and form of oxide film also varies with the composition: thus, tin oxide crystals may be present in incompletely-deoxidised tin bronze; a zinc oxide film covers the liquid surface in gun metals; and phosphorised bronzes have a highly

1 W. T. Pell-Walpole and V. Kondic. *J. Inst. Metals*, 1944, **70**, 275.

fluid liquid phosphate on the surface of the molten metal. These features may exert a markedly varying action on the ease of internal feeding in the final stages of solidification of an ingot.

The effects and methods of control of shrinkage in shaped castings have been dealt with recently by Baker.[2] The present research was undertaken to determine experimentally the effect of the various factors discussed above on the extent and form of shrinkage porosity in chill ingots, with the aim of developing a quantitative relation between size and form of ingot, and the casting procedure required to minimise harmful types of porosity in these products.

General Experimental Procedure

The general procedure adopted was to make melts from either scrap or virgin metals under degassing fluxes[3]

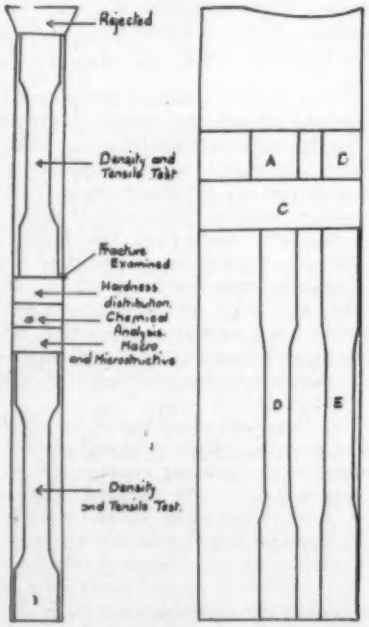


Fig. 1.—Method of sectioning 14 in. × 1 in. ingots.

Fig. 2.—Method of sectioning ingots from 3 in. dia. mould.

A—Central specimen for microstructure and hardness.
B—Edge specimen for microstructure and hardness.
C—Analysis specimen.
D—Central tensile specimen.
E—Edge tensile specimen.
Density contours at intervals of .05 grm./c.c.

containing cupric oxide, sharp sand and fused borax in the proportions 1 : 2 : 2 for virgin, 1 : 1 : 1 for scrap melts. The flux was charged with the metal into a Salamander pot in a gas-fired pit furnace. After fusion was complete sufficient time was allowed for the melt to reach the desired temperature (100° C. above the required casting temperature). The flux was then stirred into the metal, allowed to separate, thickened with dried sea sand and removed completely from the metal surface. The necessary amount of deoxidant (15% phosphor copper) was added, and other additions, including those to compensate for oxidation losses were made, prior to pouring.

Each melt was poured through a centrally-drilled, pre-heated Salamander pot used as a funnel under the selected conditions. Pouring rates were controlled by the aperture size in the Salamander funnel; temperatures of metal and mould were measured with chromel-alumel thermocouples. Details of the various casting conditions examined are given in Tables I–VIII.

Ingots were examined for surface characteristics, and then sectioned for examination of fracture, macro- and micro-structures, and for tensile and hardness tests. Details of tensile

TABLE I.

CASTING DETAILS, ANALYSES, AND HARDNESS OF BRONZES PREPARED FOR INVESTIGATION OF EFFECT OF POURING CONDITIONS ON SOUNDNESS AND PROPERTIES.

Analysis. %.				Pouring Rate. Lb./min.	Pouring Temperature. °C.	D.P. Hardness.	
Sn	P.	Zn	Pb			Centre.	Outside
10·8	0·79	—	—	4	1,300	122	122
10·7	0·85	—	—	4	1,250	123	123
10·8	0·79	—	—	4	1,200*	118	124
10·7	0·86	—	—	4	1,150*	116	118
—	—	—	—	4	1,100	Did not fill	
10·3	0·82	—	—	6	1,300	123	123
10·3	0·82	—	—	6	1,250	120	120
10·3	0·63	—	—	6	1,200	116	117
10·3	0·63	—	—	6	1,150	116	120
10·2	0·57	—	—	6	1,100*	122	120
10·7	0·83	—	—	8	1,300	119	118
10·6	0·62	—	—	8	1,250	123	122
10·4	0·88	—	—	8	1,200	120	128
10·7	0·53	—	—	8	1,150	114	118
10·9	0·61	—	—	8	1,100	122	123
11·1	0·75	—	—	16	1,300	114	115
10·4	0·88	—	—	16	1,250	116	124
10·4	0·65	—	—	16	1,200	121	120
10·4	0·77	—	—	16	1,150	119	118
10·4	0·77	—	—	16	1,100	107	123
10·5	0·90	—	—	32	1,300	120	117
10·5	0·90	—	—	32	1,250	113	124
10·7	0·90	—	—	32	1,200	117	120
10·7	0·50	—	—	32	1,150	115	117
10·7	0·71	—	—	32	1,100	104	120
8·6	0·10	—	—	8	1,300	121	125
8·7	0·11	—	—	8	1,250	130	132
9·0	0·06	—	—	8	1,200	122	128
8·8	0·08	—	—	8	1,150*	—	—
8·8	0·12	—	—	16	1,300	122	136
8·4	0·12	—	—	16	1,250	120	142
9·1	0·08	—	—	16	1,200	116	140
8·8	0·09	—	—	16	1,150	120	128
8·9	0·12	—	—	32	1,300	120	122
8·6	0·11	—	—	32	1,200	110	126
8·3	0·09	—	—	32	1,200	108	129
8·1	0·08	—	—	32	1,150	104	134
8·8	0·08	—	—	64	1,300	105	127
8·8	0·06	—	—	64	1,250	96	130
8·5	0·05	—	—	64	1,200	92	130
8·7	0·04	—	—	64	1,150	85	132
10·1	0·01	2·1	—	8	1,300	123	129
9·9	0·01	2·2	—	8	1,250	128	132
9·9	0·01	2·2	—	8	1,200	127	137
10·1	0·01	2·1	—	8	1,150*	127	132
—	—	—	—	8	1,100	Did not fill	
9·8	0·01	1·8	—	16	1,300	80	112
9·7	0·01	1·7	—	16	1,250	91	116
10·0	0·01	1·9	—	16	1,200	101	117
10·0	0·01	1·9	—	16	1,150	110	124
9·8	0·02	1·9	—	16	1,100	103	121
9·8	0·02	2·0	—	16	1,050*	103	125
9·9	0·02	2·1	—	32	1,300	81	113
9·9	0·02	2·1	—	32	1,250	80	116
9·9	0·03	1·8	—	32	1,200	88	122
9·9	0·02	1·8	—	32	1,150	83	117
9·8	0·02	2·1	—	32	1,100	100	123
9·9	0·02	2·1	—	32	1,050	105	131
5·0	0·10	5·0	5·0	64	1,200	115	103
5·0	0·10	5·0	5·0	32	1,200	87	105
5·0	0·10	5·0	5·0	16	1,200	87	112
5·0	0·10	5·0	5·0	8	1,250	91	103
5·0	0·10	5·0	5·0	8	1,200	98	105
5·0	0·10	5·0	5·0	8	1,150	119	124
Sn	P	Ni.	Pb				
4·5	0·10	0·9	18·4	32	1,200	—	—
4·5	0·10	0·9	18·4	16	1,200	—	—
4·8	0·10	1·0	19·2	12	1,200	—	—
4·8	0·10	1·0	19·2	12	1,200	—	—
4·8	0·10	1·0	19·8	12	1,250	—	—
4·8	0·10	1·0	19·6	8	1,250	—	—
4·8	0·10	1·0	19·6	8	1,200	—	—

* Mould did not fill completely.

2 W. A. Baker. "Shrinkage and Gas Effects in the Casting of Non-ferrous Metals and Alloys." Research Report. R.N.F.M.R.A. No. RRA. 661, 1944.
3 W. T. Pell-Walpole. *J. Inst. Metals*, 1944, **70**, 127.

A—Poured at 1,100° C., 16 lbs./min. B—Poured at 1,200° C., 16 lbs./min. C—Poured at 1,300°C., 16 lbs./min. D—Poured at 1,200° C., 8 lbs./min. E—Poured at 1,200° C., 32 lbs./min.

Fig. 3.—Effect of pouring conditions on macrostructure (88/10/2).

test-pieces are shown in Figs. 1 and 2; hardness tests were made on a Vickers machine with a diamond indentor and a 20 kgm. load.

Part I.
The Effects of Rate and Temperature of Pouring with Constant Mould Conditions

(*a*) *Preparation of Ingots.*—Melts of 2B8 bronze (10·5% tin, 0·75% phosphorus) were prepared from virgin metals, melts of Admiralty gun metal (88/10/2), leaded gun metal (85/5/5/5) low-phosphorus bronze (9% tin, 0·1% phosphorus) and leaded bronze (20% lead, 5% tin) from ingotted scrap. Ingots 14 in. × 1 in. dia. were cast in each composition at various rates and temperatures of pouring (see Table I) into chill moulds, pre-heated to 100° C., and dressed with tallow/charcoal. The lowest practicable pouring rate for various temperatures is shown in Table I. This minimum rate varied with alloy composition, being appreciably lower with the very fluid 2B8 bronze than with the other alloys.

(*b*) *Surface Quality of Ingots.*—In general the best surface quality was obtained with all compositions when the pouring rate was the minimum for complete filling of the mould without cold shuts, at a pouring temperature of 1,200° C. Higher rates tended to produce charcoal traps and local surface roughness, due to turbulence; this effect became very serious in the ingots of low-phosphorus bronze poured at 120 lb./min., the fastest rate investigated.

Local blowing porosity tended to occur with high pouring temperatures (1,250° C. or above), and high pouring rates (> 16 lb./min.), particularly in the high-leaded bronze, slightly less frequently in 2B8, and rarely in the low-phosphorus bronze and gun metals.

All alloys were free from tin-sweat at all rates and temperatures of pouring investigated. The ingot tops were almost flat with the minimum pouring rate, but a slight sink appeared with higher rates, deepening as either rate or temperature of pouring were increased.

(*c*) *Ingot Fractures.* Cross-sectional fractures through ingots of all compositions poured at the minimum rate, showed a fine silky texture. As the rate increased, central discolouration appeared, indicating shrinkage fissures, and at the highest rates studied, small shrinkage cavities became visible to the naked eye. For constant rate of pouring these shrinkage effects increased slightly with increase of pouring temperature, but at very low temperatures (1,100° C. or lower) and high pouring rates, spherical traps of injected air were sometimes present.

In leaded bronze and 2B8 bronze only, a sub-surface layer of very fine spherical traps was present in some ingots, probably caused by the entrapping of insoluble gases generated from the mould dressing.

All alloys showed a gradual transition from greyish-brown at the surface to red-brown at the centre. This regular variation is quite distinct from the central discolouration which indicates shrinkage fissures, and is caused by the decreasing quantity of $(\alpha+\delta)$, or of $(\alpha+\delta+Cu_3P)$ complex from the rapidly-chilled surface to the more slowly-cooled centre of the cross-section. A similar variation occurs in respect of the average tone of the whole cross-section, as temperature of pouring is increased.

(*d*) *Macrostructures.* Macrostructures show a marked increase of grain size with rise of pouring temperature, and a slight increase with increase of pouring rate (see Fig. 3). The structures are, in general, equi-axed, but columnar grain-formation occurs with the highest pouring temperature investigated (1,300° C.), irrespective of pouring rate with the low-phosphorus bronze, but only at the slowest rate with the other alloys.

(*e*) *Microstructures.* The complex $(\alpha+\delta)$ or, in the case of 2B8 bronze, $(\alpha+\delta+Cu_3P)$ decreases in amount and changes in form from intercrystalline envelopes to isolated areas as the temperature is raised, and to a less degree as the rate of pouring is increased.

In the leaded alloys, the lead distribution is finest and most uniform when pouring rate is at the minimum practicable.

Three types of micro-shrinkage porosity occur, common to all the alloys examined, and similar to those described elsewhere for strip ingots of tin bronze.[1] They vary in relative amount according to the pouring conditions employed:

1. Fine uniformly-distributed shrinkage pinholes associated with the $(\alpha+\delta)$ eutectoid (Fig. 23) occur under all conditions of pouring, increasing in size with pouring temperature, but not varying appreciably with pouring rate. These are the only type of shrinkage cavity present when the alloys are poured at the lowest possible rates.
2. Fine intercrystalline fissures (Fig. 24) occur with the more rapid rates of pouring, particularly when the pouring temperature is high. They occur more extensively in low-phosphorus bronze and in Admiralty gun metal than in the more fluid leaded or high-phosphorus bronzes.
3. Coarse axial shrinkage cavities (Fig. 25) occur in all alloys when the rate of pouring is excessive (32 lb./min. or more), and are most severe when pouring temperature is also high (1,250° C. or over).

(*f*) *Density.* The mean density of upper and lower machined test-pieces of each ingot is considered to give the best representation of the extent of shrinkage porosity, since the latter is concentrated near the central axis, while the various forms of surface and sub-surface porosity are eliminated in the preparation of the test-pieces. These values are plotted as contour maps against rate and temperature of pouring for each of the alloys studied in Figs. 4–7. For each alloy, pouring rate has a much greater effect on density than has pouring temperature, and maximum densities are obtained

with the slowest pouring rate practicable for each composition, respectively. The maxima obtained represent 0·5–1·0% porosity, as compared with ideal densities of the respective alloys. For a constant temperature—*i.e.*, any horizontal section across the contour maps (see Fig. 8), density decreases as pouring rate increases, according to the general expression

$$D = A - R^B$$

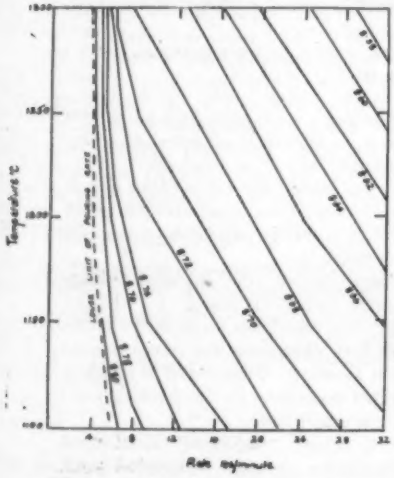
where D = density
R = pouring rate.
A and B = constants.

The value of B is greater with the high phosphorus bronze than with the other, less fluid, compositions.

Increase of pouring temperature causes a gradual but continuous decrease of density in high phosphorus bronze and in Admiralty gun metal. This effect is very slight with the lowest pouring rates, but becomes relatively important when the rate is 16–32 lb./min. In the low-phosphorus bronze, increase of pouring temperature from 1,100–1,200° C. produces an increase of density, but further increase from 1,200–1,300° C. causes a slight fall in density. The ridge of higher densities at 1,200° C. probably indicates that this is the lowest temperature at which continuous feeding can occur to the surface of the ingot before crystallisation commences in the pouring stream. The effect of pouring temperature on the density of 2B8 high-phosphorus bronze is reflected in the variation of density from bottom to top of the ingots. The upper half of each ingot is always somewhat denser than the lower half, particularly when pouring rate is low. This variation is caused by the effective fall in pouring temperature during the filling of the mould, and this fall becomes more pronounced as the pouring rate decreases.

The bulk densities of the ingots of each composition are generally somewhat higher than the test-piece densities, particularly when the pouring rate is high. This is caused by the tendency for shrinkage porosity to be concentrated along the vertical axis of the ingot. Bulk densities vary with pouring conditions in a rather less regular manner than the test-piece densities, due to the interfering effects of surface and sub-surface porosity in the former; these defects occur somewhat irregularly, but tend to be more pronounced in the lower halves of the ingots. Hence, the bulk densities show a greater variation from top to bottom of the ingot, than do the test-pieces.

Fig. 4.—Effect of pouring conditions on density of 2B8 test-pieces.



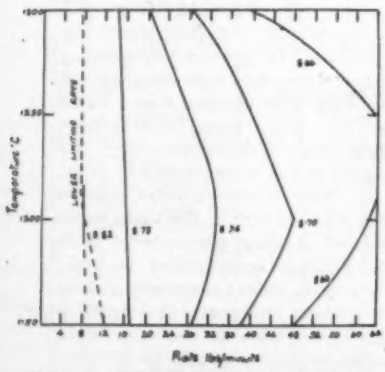


Fig. 5.—Effect of pouring conditions on density of "as-cast" low-phosphorus bronze (tin 9% phosphorus 0.1%). Contours in grms./c.c.

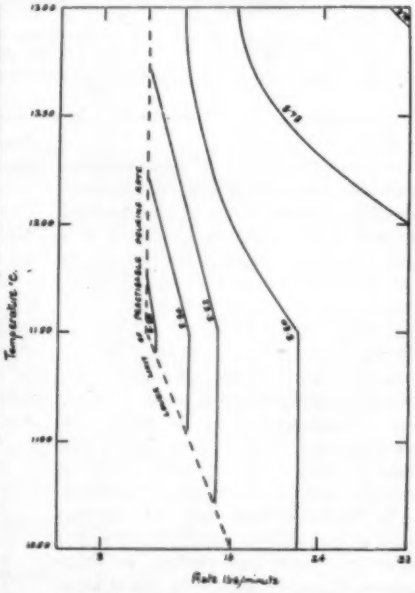


Fig. 6.—Effect of pouring conditions on density of "as-cast" 88/10/2 gun metal. Contours in grms./c.c.

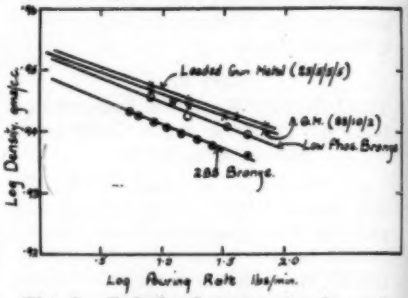


Fig. 8.—Relation between density and pouring rate.

The effect of annealing on the density of the upper test-bars was examined for the low-phosphorus bronze, and for the 88/10/2 gun metal. In both alloys there is a slight fall in density (0·02–0·05 gms./c.c.) similar to that reported previously for low-phosphorus bronzes.[4] This effect does not vary appreciably with pouring conditions.

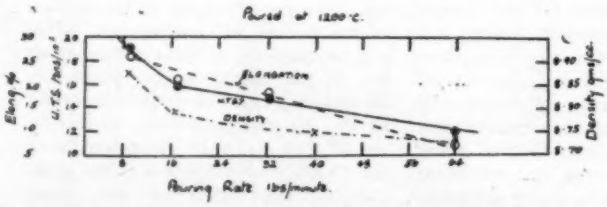


Fig. 7.—Effect of pouring rate on density and tensile properties of 85/5/5/5 gun metal chill-cast ingots (14 in. × 1 in.)

(*g*) *Tensile Properties.* The mean values of the results of tensile tests on upper and lower specimens of 2B8 bronze, and results for lower specimens only of the other alloys, all in the as-cast condition, are plotted as contour graphs with variation of rate and temperature of pouring in Figs. 7 and 9–15. With each alloy the effects of pouring rate are similar. Strength and elongation are at a maximum (in agreement with the density maxima) with the lowest practicable pouring rate. As pouring rate is increased, both values fall in a similar manner to density—*i.e.* for constant temperature,

$$M \simeq A - R^B$$

where M = Tensile strength or elongation
R = Pouring rate
A and B are constants.

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The effects of pouring temperature on strength and elongation differ appreciably in the various alloys examined, since pouring temperature affects the amount and mode of distribution of the compound phases, as well as the extent of porosity.

In general, as pouring temperature is increased the extent of porosity increases. This tends to reduce mechanical properties, but simultaneously the amounts of metastable complexes ($\alpha + \delta$, or $\alpha + \delta + Cu_3P$) formed are reduced, and their mode of distribution is changed from intercrystalline envelopes to isolated pools. These structural changes tend to improve mechanical properties. Thus, variation of pouring temperature produces two opposing effects on mechanical properties. Three cases may be distinguished: (1) these effects may tend to balance (*e.g.*, at low pouring rates none of the alloys show any marked variation of properties with variation of pouring temperature); (2) the porosity effect may increase very rapidly at high temperatures, and the compound effect at low temperatures (in this case some intermediate temperature gives optimum mechanical properties—*e.g.*, ingots of 2B8 bronze or of 88/10/2 gun metal poured at relatively rapid rates); (3) the compound effect may be much greater than the porosity effect (in this case maximum properties are obtained with the highest pouring temperatures—*e.g.*, with low-phosphorus bronze both tensile strength and elongation increase regularly as pouring temperature is increased). The effect of pouring temperature on properties is also reflected in a slight variation (2–3 tons/sq. in.) between upper and lower test-pieces of 2B8 ingots, poured slowly. These ingots take a long time to pour, and a severe fall in effective pouring temperature may occur, so that in bars poured at nominal temperatures of 1,100°–1,150° C. the metal temperature may be approaching the liquidus temperature before the ingot is filled. In bars poured more rapidly the time of filling is too short to permit any effective fall in pouring

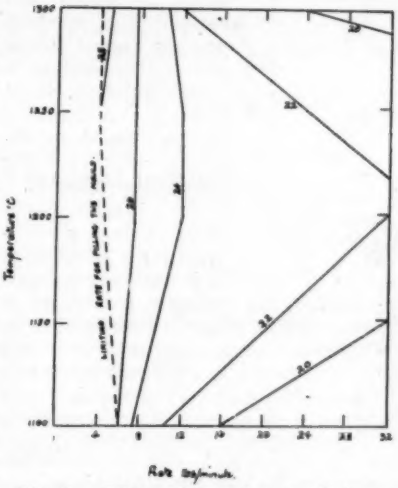

Fig. 9.—Effect of pouring conditions on tensile strength of chill-cast 2B8 bars. Contours in tons/sq. in.

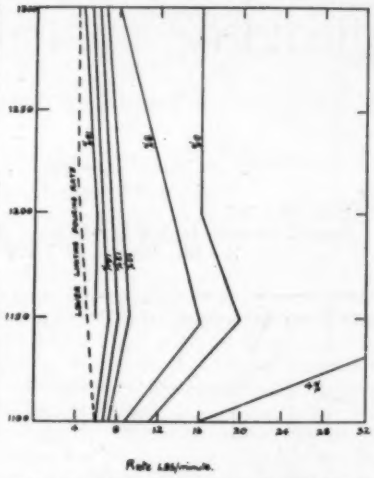

Fig. 10.—Effect of pouring conditions on elongation of 2B8 tensile test-pieces.

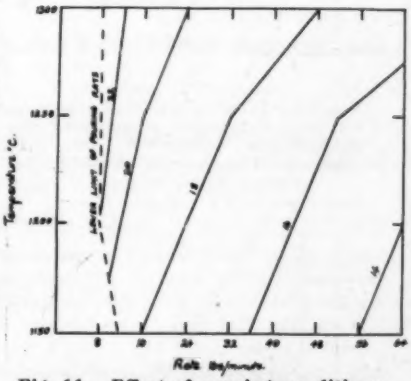

Fig. 11.—Effect of pouring conditions on tensile strength of "as-cast" low-phosphorus bronze (9% tin, 0.1% phos.). Contours in tons/sq. in.

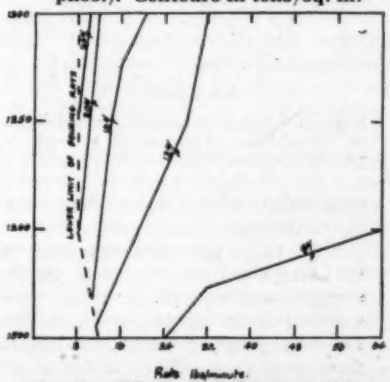

Fig. 12.—Effect of pouring conditions on elongation of "as-cast" low-phosphorus bronze (9% tin, 0.1% phos.).

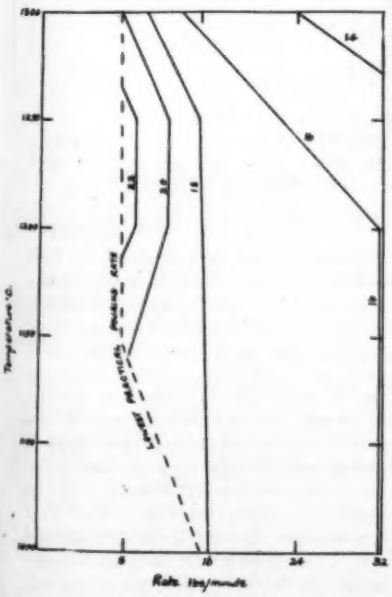

Fig. 13.—Effect of pouring conditions on tensile strength of "as-cast" 88/10/2 gun metal. Contours in tons/sq.in.

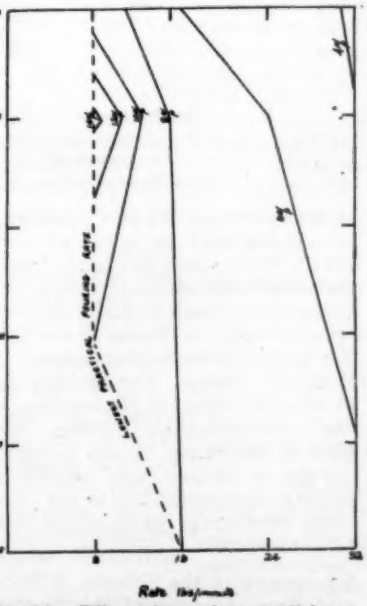

Fig. 14.—Effect of pouring conditions on elongation values of "as-cast" 88/10/2 gun metal.

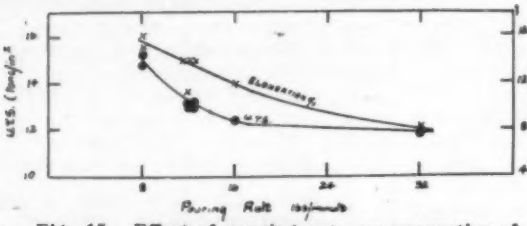

Fig. 15.—Effect of pouring rate on properties of leaded bronze ingots (20% lead, 5% tin) 14 in. × 1 in., poured at 1,200° C.

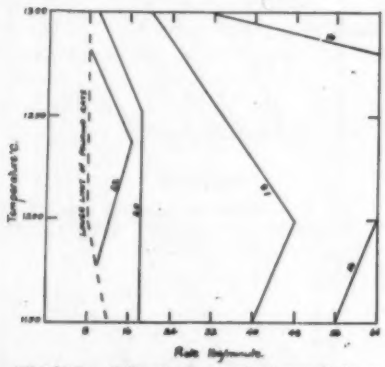

Fig. 16.—Effect of pouring conditions on tensile strength of annealed low-phosphorus bronze (9% tin, 0.1% phos.). Contours in tons/sq. in.

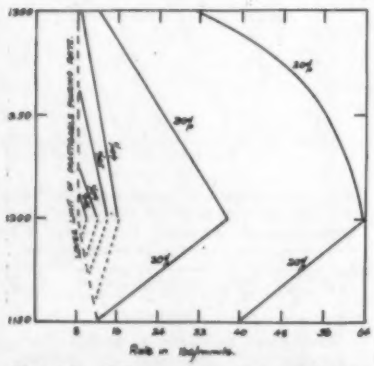

Fig. 17.—Effect of pouring conditions on elongation of annealed low-phosphorus bronze (9% tin, 0.1% phos.).

temperature from the bottom to the top of the bar.

When alloys are annealed to remove the compound phases, then tensile strength and elongation vary in a similar manner to density (Figs. 16–19)—*i.e.*, porosity alone determines these properties. Thus, in low-phosphorus bronze after annealing, optimum properties with constant pouring rate are obtained when the pouring temperature is 1,200° C. (corresponding to maximum density), while in the as-cast condition optimum properties result from pouring at 1,300° C. Similarly, with annealed 88/10/2 gun metal the optimum results are obtained (for constant pouring rate) when pouring temperature is 1,100°–1,150° C., while in the as-cast condition the best results are obtained with pouring temperature 1,250° C.

Thus, in selecting the optimum casting conditions for a given alloy, it is necessary to know whether the casting is to be used in the as-cast (duplex) condition, or in the annealed (homogeneous) condition.

(*h*) *Hardness.*—The hardness of 2B8 bronze does not vary appreciably with pouring conditions (Table I), but with 88/10/2 gun metal and low-phosphorus bronze, the hardness is highest and most uniform across the section with the lowest practical pouring rate. Increase in rate causes a marked decrease in hardness at the centres of the bars due to the accompanying increase in shrinkage porosity. Pouring temperature has little effect on the hardness of the low-phosphorus bronze, but 88/10/2 gun metal shows a steady increase in hardness with decrease in pouring temperature due to the more marked increase in the amount of hard ($\alpha + \delta$) complex.

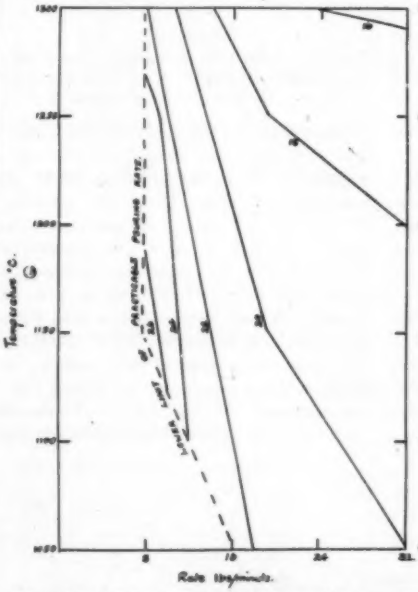

Fig. 18.—Effect of pouring conditions on tensile strength of annealed 88/10/2 gun metal. Contours in tons/sq. in.

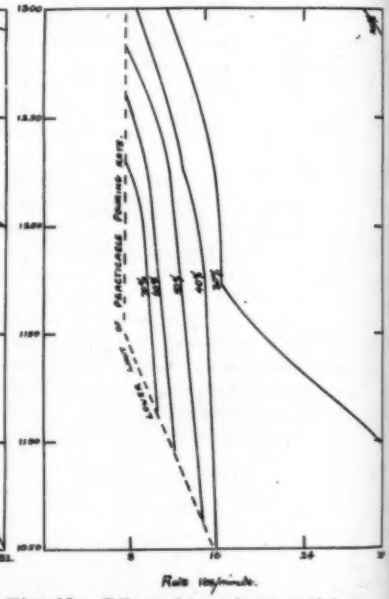

Fig. 19.—Effect of pouring conditions on elongation values of annealed 88/10/2 gun metal.

Summary of the Effects of Rate and Temperature of Pouring.

For all bronzes examined the harmful forms of shrinkage porosity can be minimised, uniform structures and optimum properties obtained by pouring at the minimum rate which will fill the mould without cold shuts. This minimum rate is lower for the more fluid high-phosphorus bronze than for the other compositions examined.

Pouring temperature has little effect on shrinkage porosity when the minimum pouring rate is used, but with rapid rates of pouring, soundness and mechanical properties diminish as pouring temperature is raised. Mechanical properties in unannealed ingots are also influenced slightly by the effect of pouring temperature on the distribution of the brittle compound phases for all rates of pouring.

(*To be continued*)

Alloy Steels for Maintenance

Although perhaps higher in first cost, alloy steels save in replacement costs and in delay time. Owing to probable need for replacement plus cost of delay due to failure, a carbon steel part may actually cost six times as much for a year's operation as the same part in alloy steel, even if the initial cost of the alloy steel part is twice that of the carbon steel part. Comparative properties and factors to consider in selecting the proper steel for a given application are discussed by J. A. Rosa* who tabulates parts used in steel plant maintenance, giving suitable types of steel for these parts and the heat treatment required.

Iron and Steel Engineering, 1945, **22**, 37-45.

Experimental Electric Arc Furnace Smelting

By H. S. Newhall

In electric smelting research, conditions, equipment, materials, temperatures, etc., should be as nearly as possible those most probably required on full scale operation. This discussion, which formed part of a symposium on "The Electric Furnace and Its Products" presented at a recent meeting of The Electrochemical Society, covers furnace design, choice of place, electrodes, refractories, sizing of charge components, and interpretation of experimental results.*

CONTINUED development in the advancement of electric furnace design and operation has opened a field which will result in a great many changes in present-day practice and the development of new operations and processes. Usually the first step in this field is research to determine the possibilities of any new ideas. After the preliminary book work has been completed, actual experimentation may be carried out to determine the economical aspects of the new process. Many of these new ideas have great merit but the economical value is oft-times lost due to improper handling of the experimental test runs in the electric furnace.

The proper type of electric furnace should be selected that will do the job. For instance, the same furnace cannot be used to obtain economical values both on operation for the smelting of tungsten ores and for the smelting of copper sulphide concentrates. These operations are completely opposite in characteristics. In the treating of tungsten ores for the production of ferrotungsten, the furnace must have a very high concentration of heat in a very small area, whereas in copper sulphide concentrate smelting for copper matte, the heat must be distributed over a large area.

To list the types of electric furnaces in groups is rather difficult because there are many variations of electric furnaces which combine the characteristics of one type of furnace with those of another. There is, for example, the comparison between the batch-type furnace and the continuous-type furnace. For the production of high-carbon ferro-alloys, pig iron, copper matte, etc., a continuously operated furnace is used in which the charged material is intermittently fed to the top of a stationary furnace and the molten product tapped through a tap hole below the level of the molten bath.

The batch- or tilting-type furnace, which tilts to empty, is generally used in processes in which the strict control of the chemical analyses is brought about by a refining period in which a solid ingredient is mixed into a molten bath.

Power Supply and Electrodes

Furnaces can further be divided into the 3-phase, 3-electrode furnaces; single-phase, 2-electrode; 3-phase, 6-electrode; and single-phase, 1-electrode furnaces. The 3-phase furnace is the most popular type furnace from many stand points. It maintains an even distribution on our accepted, 3-phase, electric power distribution systems. High electric efficiency can be obtained by closing the delta at the electrodes.

Three-phase furnaces can be further divided into those types that use amorphous carbon electrodes and those that use graphite electrodes. Normally, operations that necessitate high power input per unit area of hearth area require graphite electrodes, whereas processes necessitating the distribution of the resultant heat over a large area use carbon electrodes.

Single-phase, 2-electrode furnaces have generally passed out of the picture due to the difficulties in meeting the requirements of the power companies. However, there are a few cases where a high concentration of heat is desired and where it is not advantageous to use a single-phase, single-electrode furnace.

The 3-phase, 6-electrode furnace is used on large installations where distribution of heat is required and large electrodes are not advantageous.

The single-phase, single-electrode furnace is advantageous where a high concentration of heat necessarily eliminates the physical possibility of inserting two or three electrodes in the furnace crucible. An example of this is found in the production of ferrotungsten.

Open v. Closed Type

Another comparison can be drawn between the open-top and the closed-top type furnace. The closed-top type furnace is advantageous from the standpoint of recovery of useful waste gases. For instance, in the production of pig iron a gas of 250 B.t.u./cu. ft. is given off which can be successfully employed as a heating gas in other parts of the plant. In ferro-alloy and carbide operations, except for large installations, an open-top furnace is used because the gases have no economic value at the plant site and the maintenance of a refractory roof would be costly. It is interesting to note that in most cases where a refractory roof is used the operations are such that a heavy slag blanket, together with an overburden, protects this roof from the high temperatures.

There is considerable interest at the present time in the installation of hooded furnaces which do not use a refractory roof but a water-cooled, steel hood, maintained for the recovery of gases and for the conservation of heat to raise the temperature of the overburden and save on electric power. It is our prediction that in the future, which is becoming highly specialised to the recovery of all possible waste energy, all furnaces will sooner or later be covered to redeem these losses.

Round v. Rectangular Furnaces

There have been discussions among electric furnace designers for many years concerning the advantages and the disadvantages of locating the electrodes in the 3-phase furnace at the vertices of an equilateral triangle in a round furnace or locating the electrodes in a rectangular furnace with the 3 electrodes in line. The round furnace, with the electrodes at points of a triangle, has

* *Trans. Electrochem Soc.*, Vol **88**, 143–149.

an electric advantage because of the balanced load. The rectangular furnace, with 3 electrodes in line, has considerable mechanical advantage, from the standpoint of charging, to protect the furnace walls; it has fewer hot spots on the sidewalls, and a more symmetrical charging operation.

There are two items that seem to be the governing factors in design: the slag and the temperature of operation. In most ferroalloy operations the slag is at a minimum and the temperature is high so that a carbon lining is used. In operations such as the manufacture of pig iron, there is a heavy slag blanket and refractory sidewalls are used. In this case the troublesome factor is the heat conductance of the hot slag. A round furnace, in this case, would not be advantageous due to the tendency of this slag to transfer heat very rapidly to the side-walls and cause hot spots.

The most recent designs have combined these two designs by using a rectangular furnace, but placing the middle electrode slightly out of line, thus constructing a flat triangle to unify the electric load.

Most of the above-described furnaces produce a metallic product, together with a slag. There is an additional type of furnace which operates very similarly to a closed ferroalloy furnace, namely the volatilisation type furnace, as used in the production of *phosphorus* and *magnesium*. This type of furnace can be built on the lines of any one of the designs discussed above, the only outstanding feature being that the values are recovered from a condensed volatile given off during the reaction.

The Self-Baking Electrode

Mention has been made above concerning carbon electrodes. This closed-type group can be divided into those that use preformed carbon electrodes and those that use self-baking electrodes. Actually, as far as research is concerned, this has no great value, as preformed electrodes are used in almost all cases on small furnaces, but in discussing types of furnaces this item should be definitely brought forward.

A preformed electrode refers to an electrode manufactured in sections and shipped to the plant site to be added one-upon-the other in the furnace as they are consumed. A self-baking (Söderberg) electrode refers to a steel-encased electrode which is constructed above the furnace proper by filling a steel casing with electrode mixture and paste and having the heat of the furnace bake the electrode.

Self-baking electrodes are used in the United States on furnaces using 40 in. or larger electrodes. In other countries, where the cost of preformed electrodes is high and raw materials are available for self-baking electrodes, the size of the self-baking electrodes can be decreased considerably.

Slag Reactions

It should be kept in mind that from the chemical standpoint the electric furnace is a metallurgical tool that is similar to the reverberatory furnaces and the blast furnaces, but from an operational standpoint it is entirely different. In the electrode furnace, power is transferred into heat within the slag at the reaction zone. In other words, the heat is actually developed at the most opportune location in the furnace to promote completion of the chemical reactions. This will result in a better recovery of the materials desired. Because the heat is generated within the slag, it is possible to eliminate a large percentage of the fluxes used in the open hearth furnace to fluidise the slag. In some instances where the blast furnace necessitates the use of a 20% limestone charge, it is found that in the electric furnace we can obtain the same, and possibly better, results by using only 5% to 10% limestone. All of this results in a saving in favour of the electric furnace.

Sizing of Materials

If the operation is strictly a melting operation, such as the melting of copper sulphide concentrates into matte, the sizing of materials is of minimum importance as the volume of gases given off is small and fine materials are not objectionable. However, in a reduction operation, such as the production of pig iron, the sizing of the charged material is very important. The porosity of the overburden must be such that the gases are able to filter through the overburden without encountering large resistance. If there are a great number of fines in the material charged, the pores are closed up and the reaction gases are forced to pass through small apertures, resulting in large velocity through the overburden and thus carrying away a large amount of the fine material to the flue.

Uniform sizing is also important. Just as in any other metallurgical operation, contact between the various components of the charge should be maintained at a maximum.

The experimental electric furnace operation should be carried out in the same general type of equipment that will most probably be used in full-scale production. An 8 in. carbon electrode is the smallest threaded carbon electrode that can be purchased without special order, and therefore this limits the size of furnace that can be used in experimental electric furnace work to give comparative electrode figures. For example, many experimental electric furnace runs have been carried out in 25 and 50 kva. furnaces where it has been necessary to use graphite electrodes because carbon electrodes were not available in 3 in. and 4 in. sizes. This means than an interpretation from the consumption of graphite in the small furnace to the consumption of carbon in the large furnace has to be made. It may be possible for an experienced electric furnace operator to make estimates along this line, but it is advantageous to have actual carbon electrode results which are more valid. Along this same line, heat losses on very small furnaces are so great that most operations of 25 and 50 kva. furnaces give no comparative value whatsoever on kw-hr. consumption.

Complete notes should be kept by a competent observer on all experimental runs. There are many indications during a test run which, if noted, will help determine the correct interpretation of the test. Some of the factors which should be carefully studied in any electric furnace research are the following:—

1. Complete analysis of materials to be used.
2. All available data concerning any previous research on this type material.
3. Kw-hr. consumption.
4. Electrode consumption.
5. Condition of refractories.
6. Voltage.
7. Amperage.
8. Slug fluidity.
9. Completion of reaction, time required.
10. Metallurgical weight balance.
11. Flue gas analyses.

12. Losses due to volatilisation.
13. Mechanical losses in the slag.
14. Metallurgical losses in the slag.
15. Heat equilibrium between kw-hr. input and hearth area.
16. Sampling and analyses of products, slags and by-products.
17. Colour of flames.
18. Level of electrodes.
19. Temperatures involved.
20. Miscibility between slag and metal products.

Electric characteristics are very important. If the experimental unit used is designed properly, electric results can be very easly transferred to the design of a larger furnace. If oversize or undersize electrodes are used on small work, this will result in a change of the electric characteristics of the operation. The voltage or the amperage may not be truly indicative of that required in larger furnace operation. If the experimental hearth area is inadequate, superheating of the material results, thereby changing the normal electric characteristics of the operation involved.

As much as possible, no substitutes for ores, fluxes, or reducing materials should be used in the small-scale operation. For instance, if coal is to be used as the reducing agent for large-scale operations, coke should not be used in the experimental work. This would give rise to decided differences in operation and possibly in results, because coal is a very poor conductor of electricity whereas coke is a rather high conductor of electricity. This, of course, is an extreme case, but it is important to use the identical material on the small scale that is to be encountered in full-scale operation.

Refractories

Refractories are sometimes blamed for the incorrect design of equipment. A number of years ago experiments were carried out on nickel ores in which the slag was acid and the metal product was basic to neutral. Starting out with an acid lining it was found at the end of the first heat that about 3 in. of refractories had been eaten away around the slag line. The complete lining was taken out and a basic lining put in. After the second heat the same condition was found. Believing this to be a metallurgical reaction, the lining was again taken out and a neutral chrome lining installed. At the end of the third heat the same condition existed: that is, about 3 in. of the refractories eaten away at the slag line. It was found that this was not a chemical reaction but a purely melting reaction and that all the chrome eaten away from the sidewalls had melted and sloughed to the bottom of the furnace. In other words, anyone of the three refractories would have stood up by simply reducing the heat input per unit hearth area. Refractories in the electric furnace, as far as chemical reactions are concerned, are similar in behaviour to those of any other type of metallurgical furnace. The important thing to watch in refractories of the electric furnace is the heat transfer from the reaction zone to the refractory material. If a very high temperature of the charge is necessary, thus requiring a small hearth area, it is then compulsory to change from a refractory lining to a carbon lining.

A question that has often been asked is: "What size electric furnace is capable of giving economic results along the desired metallurgical results?" This will vary according to the type of operation involved, but for general ore smelting it is not advisable to use less than 200 kva. input. Again people ask: "Why did you pick that size?", and the answer is that an 8 in. carbon electrode is the smallest carbon electrode that can easily be obtained. Smaller carbon electrodes than this can be bought on special order, but the 8 in. is believed to be the smallest size kept in stock. There are many specific exceptions to this general specification. For example, on the treatment of chrome ores a 500 kva. unit is recommended to be installed as it is more advantageous for economic results. For ferrotungsten a 50 or 100 kva. experimental unit can be used.

Do not interpret the above discussion to intimate that it is not possible to carry out successful experiments with a furnace that is not specifically designed for each individual project. Most furnaces can be remodeled and converted into an experimental smelter by making a few changes. Furthermore, a man experienced in electric furnace research can compensate partly for the wrong type of equipment used when he interprets the results of experimental work on the basis of his past experience. Thus, in the past, considerable work has been carried out on ore smelting in a tilting-type steel-melting furnace. In order to convert this to a smelting unit, by blocking off the door, installing a down-angled spout, adding sidewall refractories to raise the roof 2 to 3 courses of brick and leaving an opening in the brick for charging. In this way the material was charged at the top and the molten material tapped from the bottom, simulating a smelting unit. As most steel furnaces are designed with a high power input, it was necessary to run the furnace for these smelting operations at reduced capacity so that the sidewalls of the furnace could withstand the heat.

Conclusions

Considerable time and money can be saved in electric furnace smelting operations by first running a series of small-batch type heats in which the general metallurgical values are established. A continuous operation should be carried out in a smelting-type furnace. Do not rush into a continuous operation until most of the metallurgical values have been obtained. A small-batch run can be carried out on 100 lb. (45 kg.) of ore in a tilting-type steel furnace at very little cost, whereas to run the smelting-type furnace continuously over a period of 24 hours is rather costly and should be carried out only after the batch heats have, within reason, determined the amounts of materials to be used.

Welded Steel Mains

A NEW research group has been formed by the British Welding Research Association to examine and overcome, if possible, the causes of failure in welded mains which are subject to the action of unpurified coal gas. This research group contains representatives of the contracting companies, the Gas Research Board, gas undertakings, steelmakers, and certain members of university staffs.

A co-ordinated programme of work has been prepared and is now under consideration. When the nature of this programme is decided, arrangements will be made for research work to be carried out in universities and elsewhere and for practical trials to be made under working conditions.

Correspondence

Recovery of Usable Metal

The Editor, METALLURGIA.

Sir,

I have read with considerable interest the article by C. F. Jinks, under the above heading, in METALLURGIA, for April, 1946, describing the use of a " concrete mixer " type of rotary furnace for metal recovery from skimmings and drosses. I use similar furnaces for this work, but I think you would be interested to know how I improved the efficiency when dealing with alloys containing volatile or readily oxidisable constituents. With the co-operation of the furnace makers, Winget, Ltd., and of The Morgan Crucible Co., Ltd., the furnace was adapted to take a retort-shaped crucible and was so converted into a rotary crucible furnace.

Rotary crucible furnace for melting alloys containing readily oxidisable constituents

My furnaces, which are oil fired, but could equally well be used with gas, are worked in a tilted position with the body and crucible at 45° to the vertical, and are fired through the top cover by stationary burners. The furnace body and crucible rotate but the cover remains stationary, the spaces between the crucible and cover and the furnace and cover serving as exhaust openings. I enclose an illustration showing a furnace in operation —the mouth of the crucible can be seen projecting through the furnace cover and also the hopper which is usually used to feed the charge into the crucible.

The continuous rotation of the furnace during melting results in the charge being turned into and under the liquid metal bath as it enters the crucible. As might be expected, this action, together with the neutral or reducing conditions inside the crucible, virtually eliminates losses from the charge by oxidation, and gives a very high metal recovery. For example, I have had yields ranging from 94% on 60/40 brass swarf with a 3% oil content and a yield of 98% on white metal dust with 70% tin content.

The rotation of the furnace also improves the thermal efficiency and the charge needs very little attention once it has been fed to the hopper. In this way I have halved my labour costs for the work normally required for charging and poking down swarf, etc. I am getting a crucible life of from 25 to 34 tons of brass swarf in a furnace of 6 to 9 cwts. brass capacity with an output of 15 to 20 cwts. ingots per day.

Yours faithfully,
H. LANDSEER BAILEY.

Ingot Metal Manufacturers,
Lanmetalle Works,
Garratt Lane,
London, S.W.18.
May 7, 1946.

We are advised by Mr. W. Kenworthy of H. H. Wardle (Metals) Ltd., that the type of furnace to which Mr. Jinks refers, was invented by Mr. H. H. Wardle, managing director of his firm, about 20 years ago. Since then several improvements have been incorporated in the design, which are claimed to have increased its efficiency.—EDITOR.

The Editor, METALLURGIA.

Sir,

The adaptation by Mr. Landseer-Bailey is certainly very ingenious, and would undoubtedly be more successful where volatile materials are being used than the open type of furnace. For light alloy drosses, however, I have found the sodium chloride-fluorspar flux to be completely successful in preventing oxidation. The open type of furnace can therefore be used with this class of material, and in consequence, more furnace capacity is available for the charge than would be the case if a crucible was being used.

With regard to the comments of Mr. Kenworthy, I can only say that my article was an attempt to describe a method of recovery which I know to be of interest to the light alloy industry. The details of operation were worked out by myself in the Aluminium Corporation's works at Dolgarrog. The furnace in question was installed by a Manchester firm.

Each section of the metal industry has its own method of scrap recovery and, when my company was faced with the problem, I obtained the most likely furnace available for the purpose. I had no knowledge of any operators using the method described, but there may be many such furnaces in operation; certainly, I made no claim to originality in this matter. I am naturally interested in any possible improvements to furnaces of this type, and would be glad to hear from Mr. Kenworthy in this connection.

Yours faithfully,
C. F. JINKS,
Technical Supervisor.

Aluminium Corporation Ltd.
Dolgarrog.
May 31, 1946.

Société Anonyme Wild-Barfield

THE above new Company has been formed in Brussels. It is controlled by Wild-Barfield Electric Furnaces, Ltd., and will handle sales and service for that Company in Belgium, Holland and Luxembourg, and will also represent their associated Company, Messrs. G. W. B. Electric Furnaces, Ltd., in the same area. The new Company will handle all the products of the parent Company and its associated Company and the full resources of their research and development departments are available when required. Interested readers in Belgium, Holland and Luxembourg should write direct to S. A. Wild-Barfield, 165, Rue Belliard, Brussels.

Attack of Various Atmospheres on Copper and Some Copper Alloys at Elevated Temperatures

By A. P. C. Hallowes,* B.Sc., A.I.M., and E. Voce,** Ph.D., M.Sc., F.I.M.

Eighteen copper base alloys and five varieties of commercial copper have been tested for resistance to intermittent oxidation and scaling at 400° C. in five different atmospheres, all derived from dry air freed from acid impurities. The work is described and conclusions resulting from the investigation are given. This presentation is based on the first part of Report R.R.A. 527 of the British Non-Ferrous Metals Research Association.

Introduction

IN connection with the investigation of copper base alloys for service at moderately elevated temperatures, resistance to oxidation and scaling in various atmospheres has been determined. The materials investigated comprised :—

(*a*) Three high-conductivity and two arsenical coppers.

(*b*) Six aluminium bronzes (ranging from 2% to 12% Al), forming a group of alloys of high scaling resistance.

(*c*) A group of alloys containing small amounts of common elements added individually to copper.

(*d*) Several other alloys used industrially at elevated temperatures.

All these materials were in the form of $\frac{5}{16}$ in. diam. rod. Most of them had been cold drawn to a 50% reduction of sectional area ; the aluminium bronzes containing 10% aluminium and above were extruded. The compositions are given in Table I, where, in addition to the copper and main alloying constituents, only the chief impurities are shown.

TABLE I.—ANALYSES OF THE MATERIALS.

Nominal Composition.	Mark	Copper, %	Alloying Constituents, %	Impurities, %	
				Iron.	Miscellaneous (>0·005).
		Group a : Coppers.			
ORC, Tough Pitch	EST	99·97	—	0·002	Oxygen 0·019
CCC, Tough Pitch	ESU	99·96	—	0·002	Oxygen 0·030
OFHC, Oxygen-free	ESV	99·98	—	0·002	Oxygen 0·0001
Arsenical, Tough Pitch	EO	99·50	As 0·38	0·0007‡	Oxygen 0·072, Ni 0·034
Deoxidised Arsenical	GHA	99·41	As 0·46	0·002	P 0·07, Ni 0·06, Pb 0·01
		Group b : Aluminium Bronzes.			
2% aluminium	DMX	97·93	Al 2·06†	0·01	—
5% aluminium	FBF	94·33	Al 5·66†	0·008	—
10% aluminium	DLH	90·20	Al 9·76†	0·039	—
10% aluminium + 2·8% iron	DMY	87·07	Al 10·13 Fe 2·80	—	—
11% aluminium	ENY	88·89	Al 11·10†	0·006	—
12% aluminium	ENZ	87·92	Al 12·06†	0·02	—
		Group c : Individual Addition Elements.			
1% silver	DUU	98·95	Ag 1·01	0·01	—
1% cadmium	DYH	98·82	Cd 1·16	0·06	—
0·5% chromium	EHX/Y	99·32	Cr 0·42	0·04	—
4% manganese	DVT	93·05	Mn 3·86	0·072	—
3% nickel	DNZ	96·91	Ni 3·02	0·01	—
0·25% phosphorus	DOG	99·78	P 0·28	0·008	—
1% silicon	DPB	99·01	Si 0·97	0·01	—
2% tin (phos. deoxid.)	DQX	98·06	Sn 1·91 P 0·007	0·007	—
3% zinc	DRO	96·77	Zn 3·22†	0·006	—
		Group d : Industrial Alloys.			
2% silicon + 1% manganese (Everdur)	DXV	96·00	Si 3·02 Mn 0·95	0·038	—
0·5% silicon + 0·75% nickel (Kuprodur)	DXW	98·36	Ni 0·89 Si 0·61	0·01	Zn (?) 0·13†
5% tin + 0·1% phosphorus	DFF	94·81	Sn 5·02 P 0·11	0·005	—

† By difference.
‡ Spectrographic determination.

Experimental Work

The eighteen materials have been tested for resistance to oxidation and scaling at 400° C. in the following atmospheres :—

1. Dry air freed from acid impurities.

The same air stream with the following additions by volume :—

2. 10% of water vapour.
3. Approximately 0·1% of sulphur dioxide.
4. 5% of sulphur dioxide.
5. Moist hydrochloric acid (approximately 0·1% HCl and 1% H_2O).

The values obtained in dry air freed from acid impurities provided a basis for comparing the scaling rates in the other atmospheres, which were designed either to accelerate the attack or in some degree to represent possible service conditions. The temperature of 400° C. was selected as being the probable maximum at which copper-base alloys might be expected to retain useful mechanical properties.

Cylindrical specimens, approximately 1·25 cms. diam. and 4·45 cms. long, giving 20 sq. cms. surface area, were finished with 000 Hubert emery paper, degreased with acetone, and weighed. Each prepared specimen was supported on a silica boat, suitably modified to ensure minimum contact, and placed in a tube furnace previously preheated to 400° C. The cylinder was removed from the furnace after a period of 5 hours in a stream of the atmosphere selected and reweighed on cooling, together with any loose scale, giving the weight increment. The detachable scale was then removed with a stiff brush and the specimen weighed alone, giving the weight loss. This cycle of operations was repeated until the attack in several consecutive cycles became constant.

The air was freed from acid impurities by passing over solid potassium hydroxide, and dried by passing over silica gel. The added impurities are expressed as volume additions to the air stream which flowed at the rate of 10 litres per hour through each furnace.

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** Formerly of the British Non-ferrous Metals Research Association. Now with the Copper Development Association.

An atmosphere of air containing 10% water vapour was obtained by passing the air stream through a water bubbler heated to 74° C.; this procedure was found by experiment to introduce the required concentration of water vapour under the existing conditions. In the tests with sulphur dioxide the required amount was passed into the air stream through flow-meters designed to provide either approximately 0·1% or 5% of the gas. Hydrochloric acid was added by passing air through a solution of the acid under carefully standardised conditions; it was determined experimentally that an air stream passed at 10 litres per hour through a mixture of 7 volumes of concentrated hydrochloric acid and 3 volumes of water became contaminated, under the existing conditions, with approximately 0·1% of the gas and 1% of water vapour when the temperature of the acid was maintained at 17° to 18° C.

Results

The results are summarised in Table II. The increment and loss rates are expressed in milligrammes per square decimetre of surface area after a constant rate of attack had been attained. The increment rate represents the total increase of weight per cycle, including any detachable scale, and indicates the average rate at which oxygen or other reactive gas was being absorbed during the final stages of the test. The loss rate refers to the specimen alone, and defines its average decrease of weight per cycle after detachable scale had been removed by brushing; it indicates approximately the amount of metal removed per cycle after attainment of a steady rate of attack. If a material forms an adherent scale which either fails to exfoliate or for which any loss by exfoliation is less than the increment due to oxidation, the result will be an increase in weight of the specimen even after removal of any loose scale by brushing. In such a case the loss rate will be recorded as a negative quantity. In Table II a negative loss rate, therefore, indicates that the scale was adherent and probably protective.

The results of tests continued for a considerable number of cycles showed that exfoliative materials reached a steady rate of attack earlier than those which formed totally or partially adherent scales. The five coppers and many of the alloys were found to scale exfoliatively at a uniform rate from the initial stages of treatment, and extended testing was therefore considered to be unnecessary for such materials; the number of cycles required to establish a constant scaling rate for this type of attack varied from four to six. For the aluminium bronzes, however, and alloys offering a high resistance, ten, twenty, or even thirty cycles were necessary before an ultimate rate of scaling could be calculated from at least six consistent cycles.

TABLE III.—Dry Air Freed From Acid Impurities.

Material.	Thickness of Metal Removed (see p. 97).
12% aluminium	nil
11% aluminium	nil
10% aluminium + 2·8% iron	nil
10% aluminium	nil
5% aluminium	nil
2% aluminium	nil
3% silicon + 1% manganese (Everdur)	nil
1% silicon	0·1
0·5% silicon + 0·75% nickel (Kuprodur)	0·2
1% cadmium	0·4
5% tin + 0·1% phosphorus	0·4
3% zinc	0·4
2% tin (phosphorus deoxidised)	0·5
Deoxidised arsenical copper	0·6
Tough pitch arsenical copper	0·7
4% manganese	0·8
0·25% phosphorus	0·9
3% nickel	0·9
0·5% chromium	1·1
OFHC copper	1·2
CCC copper	1·5
ORC copper	1·5
1% silver	1·5

TABLE II.—Oxidation and Scaling at 400° C.

Material. (Nominal composition.)	Mark.	Dry Air.		Air + 10% H_2O.		Dry Air + 0·1% SO_2.		Dry Air + 5% SO_2.		Moist Air + 0·1% HCl.		
		Increment Rate.	Loss Rate.	Increment Rate.	Loss Rate	Increment Rate	Loss Rate.	Increment Rate.	Loss Rate.	Loss After Initial 5-Hour Treatment.	Increment Rate.	Loss Rate.
Group a: Coppers.												
ORC, Tough Pitch	EST	54	350	38	230	130	290	280	230	5,800	—	—
CCC, Tough Pitch	ESU	42	340	37	240	130	300	220	250	7,400	—	—
OFHC, Oxygen-free	ESV	38	280	30	210	94	220	220	260	6,200	—	—
Arsenical, Tough Pitch	EO	30	180	22	130	30	190	290	280	6,400	—	—
Deoxidised Arsenical	GHA	25	140	17	110	35	190	280	340	6,100	—	—
Group b: Aluminium Bronzes.												
2% aluminium	DMX	4	−4	7	−7	22	72	660	500	770	230	1,200
5% aluminium	FBF	0·5	−0·5	1	−1	2	−2	1,800	2,600	270	65	330
10% aluminium	DLH	nil	nil	nil	nil	3	−3	320	190	130	150	470
10% aluminium + 2·8% iron	DMY	nil	nil	nil	nil	2	−3	4	−1	130	85	240
11% aluminium	ENY	nil	nil	nil	nil	2	−2	310	120	280	120	150
12% aluminium	ENZ	nil	nil	nil	nil	nil	nil	9	6	200	130	210
Group c: Individual Addition Elements.												
1% silver	DUU	56	350	52	280	130	410	430	320	7,200	—	—
2% aluminium	DMX	4	−4	7	−7	22	72	660	500	770	230	1,200
1% cadmium	DYH	10	80	14	110	25	180	180	160	1,300	—	—
0·5% chromium	EHX/Y	40	240	39	260	31	220	380	300	5,700	—	—
4% manganese	DVT	26	180	17	80	100	250	320	240	8,100	—	—
3% nickel	DNZ	44	210	33	150	110	290	350	310	5,200	—	—
0·25% phosphorus	DOG	38	210	24	140	28	170	240	210	4,700	—	—
1% silicon	DPB	5	22	19	50	48	160	780	880	550	600	3,800
2% tin (phos. deoxid.)	DQX	36	110	31	220	42	300	630	600	10,000	—	—
3% zinc	DRO	22	100	15	90	32	250	220	160	38,000	—	—
(See also arsenical coppers in *Group a*)												
Group d: Industrial Alloys.												
3% silicon + 1% manganese (Everdur)	DXV	2	1	0·6	−0·6	1	−0·5	270	270	−30	330	2,200
0·5% silicon + 0·75% nickel (Kuprodur)	DXW	18	80	13	45	75	90	410	430	2,200	700	4,100
5% tin + 0·1% phosphorus	DFF	34	81	31	160	120	350	580	600	11,000	—	—

Increment rate represents weight increment per cycle for specimen and scale. Loss rate represents loss of weight per cycle of the specimen only, after brushing. (Negative loss rate indicates an adherent scale.) } in milligrammes per square decimetre of surface area of specimen.

When small proportions of hydrochloric acid and water vapour were mixed with the air stream, the rate of scaling was accelerated to such an extent that it was possible to select the materials showing useful resistance to this atmosphere after one 5-hour treatment. Increment and loss rates are therefore only given in this case for the nine alloys showing least attack. The gravimetric results for a single 5-hour treatment are included in Table II, and show that the remaining fifteen materials were entirely unsuitable for use in moist atmospheres containing small amounts of hydrochloric acid.

The results are considered separately in Tables III to VII, and show, in decreasing order, the relative resistance of the materials to attack in each atmosphere. Loss rates only are shown and expressed in practical terms of the thickness* of metal which would be removed from the surface of the material during ten 5-hour cycles, after a steady rate of scaling had been attained, assuming the attack to be uniformly distributed over the surface of the specimen. The thickness is given in thousandths of an inch and is an approximate value derived from the loss rate in Table II.

The aluminium bronzes and silicon alloys offered the greatest scaling resistance to dry air.

Except in the case of the 2% alloy, the aluminium bronzes formed a protective adherent scale which failed to exfoliate even after prolonged testing. The amount of detachable scale on the 2% alloy was so small, however, that the net result was a negative loss rate, since the gain in weight during each cycle was in excess of any loss by exfoliation.

The best of the silicon alloys was "Everdur," which was similar to the 2% aluminium bronze in that the exfoliations were only partial and very slight. The attack on the 1% silicon alloy, although small, was rather irregular, and is shown in Fig. 1. An appreciable loss took place during the first three cycles, but this became almost negligible during the following six; at the tenth cycle a further large exfoliation occurred before the material finally assumed a steady scaling rate until the eighteenth cycle, after which the test was discontinued. "Kuprodur" scaled at a slightly higher rate, but is in the range of useful materials for service at elevated temperatures in dry air.

Of the remaining alloys with individual addition elements, those containing small amounts of cadmium, zinc and tin showed the best resistance to scaling. The chromium copper, commercial coppers (with the exception of the arsenical which were relatively resistant) and the 1% silver alloy suffered the highest scaling losses.

The order of resistance of the materials was much the same in moist air as in dry air. The coppers, including the arsenical varieties, and the alloys containing manganese, zinc, phosphorus, nickel or silver, together with "Everdur" and "Kuprodur" were attacked to a slightly less extent, probably due to dilution of the atmospheric oxygen by water vapour. The loss rates of the 2% and 5% tin bronzes, on the other hand, were approximately doubled while the 1% silicon and the cadmium and chromium alloys also suffered slightly increased attack. The addition of water vapour had no discernible effect on the attack upon the aluminium bronzes.

From the figures given in Table IV, it is seen that "Everdur," like the aluminium bronzes, showed no attack in moist air while the other silicon alloys were nearly as good. The arsenical coppers were more resistant than the non-arsenical, the 4% manganese alloy occupied a relatively high place in the list, while zinc and cadmium also conferred good scaling resistance.

TABLE IV.

AIR FREED FROM ACID IMPURITIES AND CONTAINING 10% OF WATER VAPOUR.

Material.	Thickness of Metal Removed (see bottom previous column).
12% aluminium	nil
11% aluminium	nil
10% aluminium + 2·8% iron	nil
10% aluminium	nil
5% aluminium	nil
2% aluminium	nil
3% silicon + 1% manganese (Everdur)	nil
0·5% silicon + 0·75% nickel (Kuprodur)	0·2
1% silicon	0·2
4% manganese	0·4
3% zinc	0·4
Deoxidised arsenical copper	0·5
1% cadmium	0·5
Tough pitch arsenical copper	0·6
0·25% phosphorus	0·6
3% nickel	0·7
5% tin + 0·1% phosphorus	0·7
OFHC copper	0·9
ORC copper	1·0
2% tin (phosphorus deoxidised)	1·0
CCC copper	1·1
0·5% chromium	1·2
1% silver	1·2

Except in the case of three of the coppers, the chromium alloy, the aluminium bronzes and "Everdur," the inclusion of small amounts of sulphur dioxide had a marked accelerating effect on the scaling losses of all materials. The coppers (with the exception of the arsenical) and the chromium copper, on the other hand, although showing increased increment rates in line with the other materials, did not exfoliate to quite the same extent as in dry purified air.

TABLE V.

DRY AIR CONTAINING APPROXIMATELY 0·1% OF SULPHUR DIOXIDE.

Material.	Thickness of Metal Removed (see bottom previous column).
12% aluminium	nil
11% aluminium	nil
10% aluminium + 2·8% iron	nil
10% aluminium	nil
5% aluminium	nil
3% silicon + 1% manganese (Everdur)	nil
2% aluminium	0·3
0·5% silicon + 0·75% nickel (Kuprodur)	0·4
1% silicon	0·7
0·25% phosphorus	0·8
1% cadmium	0·8
Tough pitch arsenical copper	0·8
Deoxidised arsenical copper	0·8
OFHC copper	1·0
0·5% chromium	1·0
1% manganese	1·1
3% zinc	1·1
3% nickel	1·3
ORC copper	1·3
2% tin (phosphorus deoxidised)	1·3
CCC copper	1·3
5% tin + 0·1% phosphorus	1·5
1% silver	1·8

The aluminium bronzes and "Everdur" were the only materials showing useful resistance to this atmosphere. Although the 2% aluminium alloy formed a completely exfoliative scale from the beginning of treatment, the remaining aluminium bronzes were protected in each case by a totally adherent scale. "Everdur" possessed a very high resistance, but "Kuprodur" and the 1% silicon alloy were heavily exfoliative compared with their behaviour in dry air. The scaling loss of the 1% silicon alloy was irregular, and is shown in Fig. 1. As in the case of dry air, the scale exfoliated in relatively large amounts at intervals, but

* Assume a loss rate of W mgm. per sq. dm. per cycle, or $W \times 10^{-3}$ gms. per sq. cm. per cycle. Taking the density of copper, namely 8·9 gms. per c.c., as applicable to all materials with sufficient accuracy for the present purpose, this becomes $\frac{W \times 10^{-3}}{8 \cdot 9}$ c.c. per sq. cm. per cycle, i.e., cms. thickness removed per cycle. This is equivalent to $\frac{W \times 10^{-3}}{8 \cdot 9 \times 2 \cdot 54}$ thousandths of an inch per 10 cycles = 0·0044 W thousandths of an inch per 10 cycles.

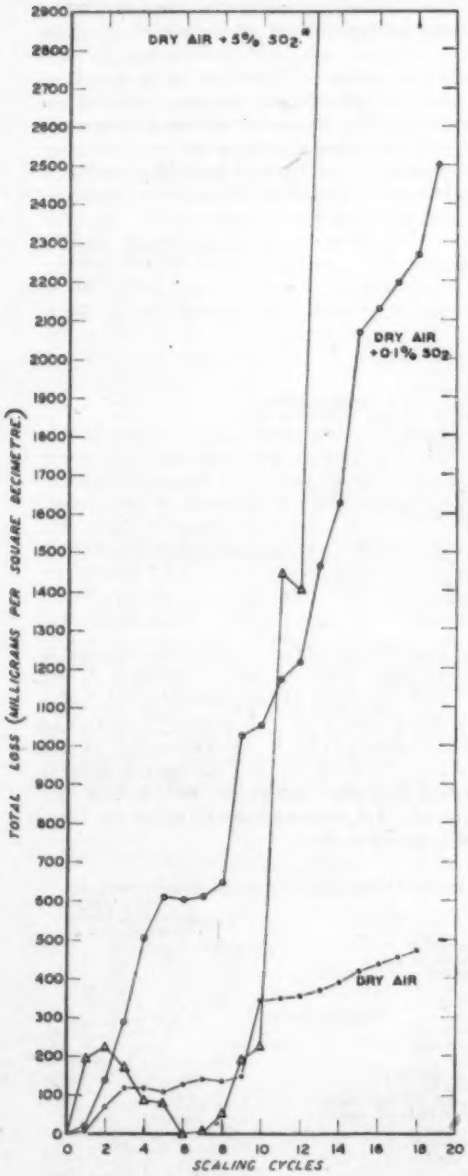

Fig. 1.—Scaling of 1% silicon alloy in various atmospheres.
* The ultimate scaling rate was established after 16 cycles.

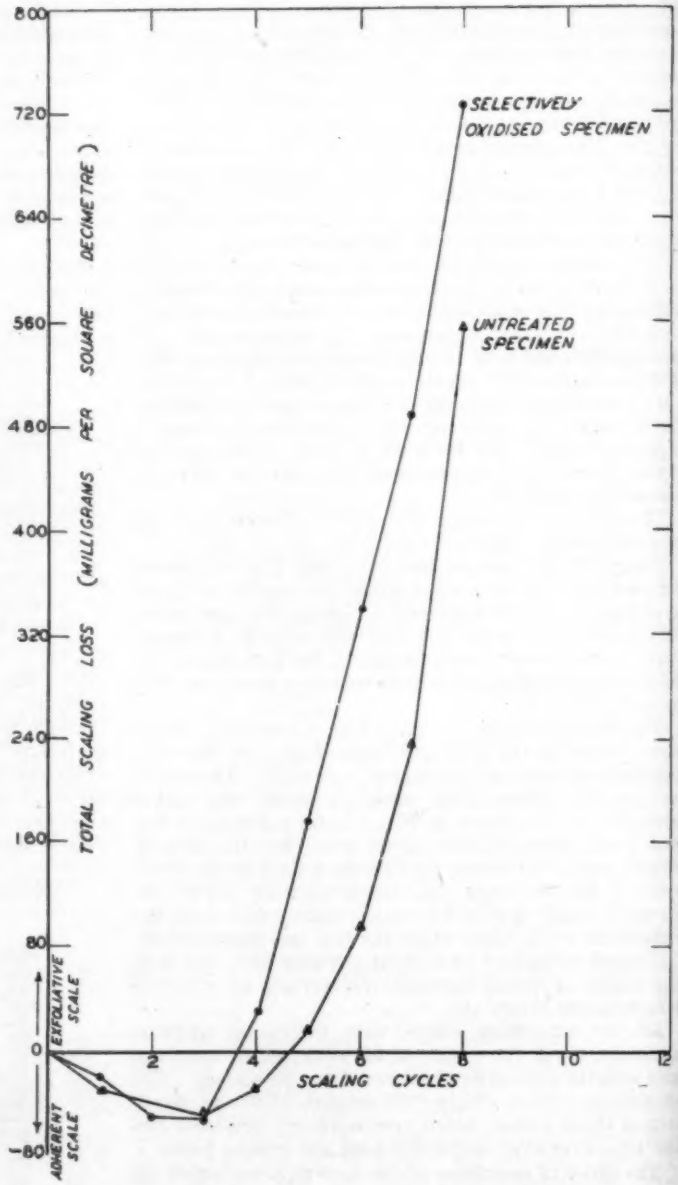

Fig. 3.—Scaling of selectively oxidised 5% aluminium bronze in an atmosphere of dry air containing 5% of sulphur dioxide.

nevertheless an average loss rate could be assessed over nineteen cycles.

Inferior scaling resistance was shown by all the alloys in *Group C* containing small amounts of individual elements. Within this group, however, the 2% aluminium, 1% silicon and 1% cadmium alloys showed the least attack in this as in the other atmospheres.

Heaviest scaling occurred in the cases of the ORC and CCC coppers, the tin bronzes and the 1% silver alloy.

The only material completely resisting this highly sulphurous atmosphere was 10% aluminium bronze containing 2·8% of iron, the exfoliations even after prolonged testing (twenty-seven cycles) remaining insufficiently large to impart a positive loss rate to the material. A white layer was formed on the surface offering protection from further attack.

Other aluminium bronzes, except the 2% alloy, built up a similar white layer which afforded protection during the early cycles. At a later stage in the treatment, however, preferentially attacked spots or wart-like excrescences were formed, fewer in number as the aluminium content of the material increased; they

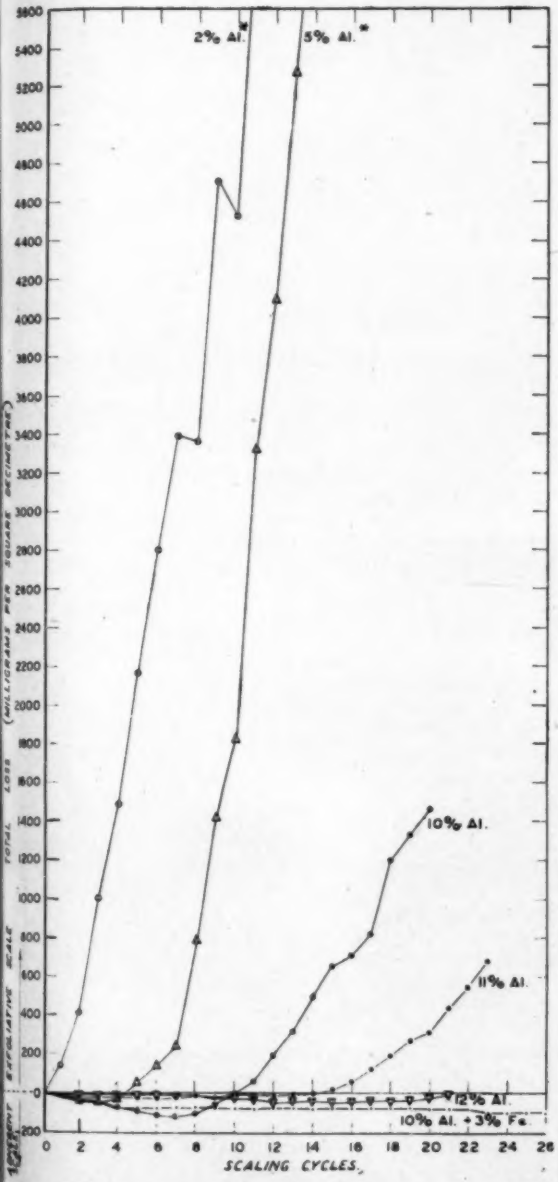

Fig. 2.—Scaling of aluminium bronzes in atmosphere of dry air containing 5% of sulphur dioxide.

*** The ultimate scaling rates of the 2% and 5% alloys were established after 15 and 22 cycles respectively.**

exfoliated during later cycles, leaving areas of a coppery appearance. The localised attack occurred over smaller areas as the aluminium content increased beyond 5%, which alloy showed the heaviest scaling of all materials in this atmosphere. The scaling losses (Fig. 2) show that exfoliation commenced at successively later stages with increasing aluminium content, but a steady rate of attack was attained after about twenty cycles. A similar localised attack was not shown by the 2% aluminium alloy, which formed a completely detachable scale from the first cycle. The loss increased slightly during later stages of the treatment,† and the scale finally exfoliated after alternate cycles, producing an average loss rate considerably lower than that of the 5% alloy. The aluminium alloys may thus be regarded as comprising two types in respect of their behaviour in this atmosphere. With 5% aluminium and above exfoliation spread from localised areas, leading in the case of the 5% alloy to very severe attack. The 2% alloy on the other hand exfoliated uniformly over its entire surface, and exhibited a scaling rate low in comparison with the 5% alloy.

TABLE VI.—Dry Air Containing 5% of Sulphur Dioxide.

Material.	Thickness of Metal Removed (see p. 97).
10% aluminium + 2·8% iron	nil
12% aluminium	nil
11% aluminium	0·7
1% cadmium	0·7
3% zinc	0·7
10% aluminium	0·8
0·25% phosphorus	0·9
ORC copper	1·0
4% manganese	1·1
CCC copper	1·1
OFHC copper	1·1
3% silicon + 1% manganese (Everdur)	1·2
Tough pitch arsenical copper	1·2
0·5% chromium	1·3
3% nickel	1·4
1% silver	1·4
Deoxidised arsenical copper	1·5
0·5% silicon + 0·75% nickel (Kuprodur)	1·9
2% aluminium	2·2
2% tin (phosphorus deoxidised)	2·6
5% tin. + 0·1% phosphorus	2·6
1% silicon	3·9
5% aluminium	11·4

The silicon alloys offered a much lower resistance to this atmosphere than to dry air, moist air, or dry air containing 0·1% of sulphur dioxide. They became covered with a white protective layer in the early stages, but this was gradually detached during later cycles, when considerably increased losses occurred, placing them among the least resistant materials. This was most marked in the case of the 1% alloy (Fig. 1), although "Kuprodur" scaled in the same way and exfoliated in large amounts after the preliminary cycles.

TABLE VII.

Moist Air Containing Approximately 0·1% of Hydrochloric Acid.

Material.	Thickness of Metal Removed (see p. 97).
11% aluminium	0·7
12% aluminium	0·9
10% aluminium + 2·8% iron	1·1
5% aluminium	1·5
10% aluminium	2·1
2% aluminium	5·3
3% silicon + 1% manganese (Everdur)	10·1
1% silicon	15·4
0·5% silicon + 0·75% nickel (Kuprodur)	18·0

Remaining materials very heavily attacked in this atmosphere.

In every case the increment rate was greater with 5% than with 0·1% of sulphur dioxide in the air stream, probably due to the formation of a larger proportion of sulphur compounds in the scale for a given quantity of metal attacked. In a few instances, namely, the ORC and CCC coppers, the 1% cadmium, 3% zinc, and 1% silver alloys, the loss rates were slightly diminished, while that of the 4% manganese alloy remained substantially unchanged.

It has already been pointed out that the presence of a small amount of hydrochloric acid with water vapour accelerated the rate of oxidation of the materials to an extent which rendered many of them entirely unsuitable for use in such an atmosphere. The scaling rates were determined only for the more resistant alloys, namely,

† Fig. 2 does not include these later stages of treatment on the 2% and 5% aluminium alloys.

those listed in Table VII, the remainder showing such heavy attack during the first cycle that further testing was abandoned.

Only the aluminium bronzes and, in much less degree, the silicon alloys offered useful resistance. The comparative superiority of the 1% cadmium alloy in the initial 5-hour treatment (Table II) did not persist in later cycles. The scales on all the materials were exfoliative from the commencement of the tests and the loss rates became regular after two and in some cases three cycles.

Effect of Previous Selective Oxidation.

The 5% aluminium bronze upon which a protective film had been formed by the selective oxidation method due to Price and Thomas* was tested for resistance to the atmospheres which had been found to produce the most extensive attack on the untreated material—namely, dry air containing 5% of sulphur dioxide and moist air containing approximately 0·1% of hydrochloric acid. Small cylindrical specimens, of the dimensions already given on page 3, were again used; they were finished on 000 Hubert emery paper and degreased with acetone. After a few preliminary experiments, satisfactory conditions for selective oxidation were obtained, and specimens so treated were practically immune from ordinary atmospheric oxidation at temperatures up to at least 800° C. under tests lasting for 4 hours.

The comparative resistance to scaling of treated and untreated specimens was determined at 400° C. during a series of 5-hour cycles in the atmospheres named.

The scaling losses obtained in dry air containing 5% of sulphur dioxide are shown graphically in Fig. 3. During the first three cycles the specimens behaved similarly and built up a thin film of tarnish, though the selectively oxidised material remained rather brighter in appearance than the untreated alloy. On the third cycle both specimens showed slight exfoliation, which continued to increase rapidly during later cycles, the selectively oxidised alloy showing the higher rate of attack. This continued until the eighth cycle, by which time the film of selective oxidation had been completely destroyed and the scaling losses of both specimens had become almost equal. The loss suffered by the untreated alloy confirmed the value given in Table II, and exfoliation tended to spread, as before, from isolated surface spots which had shown preferential attack during the initial cycles. In the case of the selectively oxidised alloy, loss by exfoliation appeared to commence, and to spread from minute areas of attack over the entire surface of the specimen. Slight surface imperfections, where the film had perhaps been inadequately or weakly formed, may have been responsible for this.

The experiment showed that the selective oxidation treatment conferred no protective properties on the alloy as far as scaling resistance to this atmosphere was concerned.

Similar experiments in moist air containing approximately 0·1% of hydrochloric acid showed that the selectively oxidised alloy exfoliated from the first cycle and at the third cycle was being attacked at the same rate as the untreated material. The tests were discontinued at this stage, since it was clear that selective oxidation again provided no protection.

Experiments at elevated temperatures in air using specimens upon which the film had been locally damaged by scratching revealed normal oxidation of the unprotected metal thus exposed.

* *J. Inst. Metals*, 1938, **62**, 21.

Summary and Conclusions

Eighteen copper-base alloys and five varieties of commercial copper (see Table I) have been tested for resistance to intermittent oxidation and scaling at 400° C. in five different atmospheres, all derived from dry air freed from acid impurities.

1. Except silver, all the additions to copper effected an improvement to a greater or lesser extent in its resistance to dry air. Silicon conferred considerable protection, but the aluminium bronzes were the most resistant materials, being almost immune from attack under these conditions.

2. Introduction of 10% of water vapour into the air stream made little difference to the scaling rates, but some of the materials showed a slightly decreased attack. The attack upon the tin bronzes, however, was increased.

3. The presence of approximately 0·1% of sulphur dioxide in the air generally increased the attack. The most resistant materials were the aluminium bronzes containing 5% or more of aluminium, and "Everdur."

4. Increase of the sulphur dioxide content of the air stream to 5% considerably increased the attack upon nearly all the materials. The greatest resistance was afforded by the 10% aluminium bronze containing 2·8% of iron and the straight 12% aluminium bronze, but this atmosphere was detrimental to the aluminium bronzes of lower aluminium content and to the 1% silicon alloy.

5. The presence of small amounts of moist hydrochloric acid in the air stream had a marked accelerating effect on the scaling of all the materials, and only the aluminium bronzes and silicon alloys showed moderate resistance to this atmosphere.

6. Selective oxidation by the method of Price and Thomas protected 5% aluminium bronze from ordinary atmospheric oxidation at temperatures up to at least 800° C., provided that the film was not scratched or otherwise damaged. It did not, however, confer any protection against the atmospheres containing sulphur dioxide or hydrochloric acid.

7. Considering the results as a whole, the arsenical coppers were less attacked than the non-arsenical, and the other additions to copper, with the exception of silver, effected an improvement to a greater or lesser extent. The presence of aluminium conferred the highest resistance and the silicon alloys gave generally good results.

In the next issue the authors will deal with the second part of the Report on the subject of "Attack of Various Superheated Steam Atmospheres upon Aluminium Bronze Alloys."

Acknowledgment

The authors are indebted to the Director and Council of the British Non-Ferrous Metals Research Association for permission to publish this paper.

MESSRS. A. C. WICKMAN, LTD., Coventry, announce that they have been appointed Sole Agents and Engineering Representatives throughout the world for the 10 Gramme Hand Operated Injection Moulding Machine, designed and manufactured by The Nene Valley Tool Co., Ltd., The Square, Earls Barton, Northampton. This Company are also developing Power Operated Machines of 1 oz. and 2 oz. capacities.

Atomic-Bomb Engineering

By A. C. Klein, F.A.S.M.E.

Production of nuclear energy on a vast scale has focussed attention on the work of scientists and has brought to scientists highly deserved recognition. But in the atomic bomb project the construction of the necessary plant was an engineering achievement of the first magnitude, and as was pointed out recently by Mr. A. C. Klein, the work of scientists and engineers, as in many other fields, was complementary, involving team work of a high order. His views on the subjects given here, although substantially condensed, indicate that the accomplishments of engineers in this project are no less than those of the scientists.*

SCIENTISTS developed the principles for separating isotopes and discovered the phenomenon of uranium 235 fission through observing the action of submicroscopic quantities of uranium. Engineers multiplied their experiments a billionfold, produced the atomic bomb, and opened a door to the era of atomic power.

In the atomic-bomb project much of the work of the engineers was carried on at the same time as that of the scientists, rather than sequentially as is usually the case. This made engineering progress more difficult, as the time schedule set up for construction frequently required decisions before the scientists were ready to make them. How well the engineers toiled is now well known.

The atomic-bomb programme could be divided into the following principal parts :—

1. Research and scientific development.
2. Design and construction of three major plants utilising, respectively, the diffusion process, the electromagnetic process and the plutonium process.
3. The operation of these plants to produce fission materials.
4. Development of a bomb to explode these fission materials.
5. Accessory developments, including the townsite construction, the development of the raw-material programme and the construction of heavy-water plants.

The contribution of one group of engineers was concentrated principally on the construction of the plant in which isotope 235 is separated from uranium by the electromagnetic or mass spectroscope method. In principle, this method is very simple. A suitable compound of uranium is first vaporised, then it is ionised or electrified by passage through an electric arc. Next, electrified atoms are given a forward acceleration by passing through electric fields having an increase in potential. Following this, the atoms enter a strong magnetic field which curves the electrified atoms in a circular path. The influence of the magnetic field is to bend the mass 235 atoms into a circle of slightly smaller radius than those of mass 238. At the end of the circle separate containers are provided for the 235 atoms and for the 238 ones.

The engineering work of this was co-ordinated closely with the University of California physicists, who were responsible for the research phase of the project. Another group of engineers also worked actively on the basic designs of the plutonium process. They assisted the metallurgical laboratory in the erection of the experimental plant. In that plant the chain-reaction process was first demonstrated. They also selected a site for a plutonium pilot plant and designed and erected the pilot-plant structures.

* In a paper presented at a Boston Section of the American Society of Mechanical Engineers. *Mech. Eng.*, 1946, **68** pp. 297-301.

After the electromagnetic plant was placed in operation and the production of atomic fission material had commenced, it was necessary to develop means of controlling atomic energy and for releasing it as an explosion over the target. The early completion of the electromagnetic plant gave the first tangible quantities of fission material with which to work out methods of handling and detonating it. By starting production early in 1944, a year earlier than any other plant, the Los Alamos group was enabled to complete the construction of the bombs which ended the war.

Another phase of the project was the design and construction of the first catalytic heavy-water plant. Heavy water is water in which the hydrogen atom has a weight of 2 instead of its more usual weight of 1. Prior to 1942, small quantities were being separated in a semipilot plant. To provide for the expected large requirements, arrangements were made for a production-scale plant to be constructed at Trail, British Columbia, as large supplies of electrolytic hydrogen existed there. Construction was started in November, 1942, and operation began in May, 1943, about eight months ahead of the first plant to be erected in the United States.

Details of the Electromagnetic Process

While it is not possible to touch on all of the important engineering features of the electromagnetic process, some of them, which can now be divulged, may be of interest.

Vacuum.—The beam of electrified uranium atoms, which is curved by the action of the strong magnetic field, must operate in an extremely high vacuum. As much air as possible must be exhausted from the containers, otherwise ionised uranium atoms would collide with atoms of oxygen and nitrogen. In so doing, they might lose their electric charges and then would no longer be subject to the influence of the magnetic field ; or else they might be deflected by the collision and would not find their way into the proper collection receptacle.

In steam-turbine practice, 1 in. of mercury, or 25 mm., is considered a high vacuum, but even such a vacuum would contain enough air to prevent any separation of the 235 isotope. The vacuum stipulated for proper separation is of the order of 10^{-6} of a millimetre or 25,000,000 times that which is standard in power-plant practice. To produce such a vacuum, enormous pumps of the diffusion type had to be designed.

Electrical Conductors of Silver.—Another interesting engineering feature was the use of silver instead of copper for electrical conductors. In all, 14,000 tons of silver were used. It is interesting to note that the value of this silver just about equals all other expenditures for the entire electrogmagnetic project. The silver was delivered in ingots to copper rolling mills, where it was

rolled into strips. These were later welded together with silver solder into conductors of the requisite length.

Production Magnets.—A third engineering feature was the design and construction of the production magnets. These are huge structures 250 ft. long, each of which contains thousands of tons of steel of the highest possible permeability. One of the problems in connection with the magnet design was that of carrying on operations in the strong magnetic field which surrounded each of the magnets. This magnetic force was so strong that when a tenpenny nail was held in the hand, a strong effort was necessary to prevent the wrist from being twisted. The pull on the nails in the heels of a pair of shoes was strong enough to make walking difficult. An ordinary wrench or piece of pipe would either be wrested from a workman's hand or if he held on to it, would be drawn against the magnet face and his knuckles skinned. To counteract this, all movable equipment and structures that came within the range of the magnetic field had to be built of non-ferrous metals or of non-magnetic steel. This made it impossible to use standard designs for any of this equipment. Large quantities of stainless and other non-magnetic steels had to be procured at a time when they were badly needed for other parts of the war programme. A complete set of small tools of beryllium copper had to be designed and produced.

Operation and Maintenance Problems.—There were other engineering problems introduced by the need of delivering a plant which could operate continuously for 24 hours per day and 365 days per year. These involved the introduction of means for water, air and oil cooling of various parts of the magnets ; development of quick-detachable pipe and electrical connections ; development of equipment for quickly detecting leaks even of microscopic size into the vacuum chambers ; provision for maintaining coolant solutions at specified dielectric strength, and servicing and repair facilities for the principal operating units involved in the process.

Process of Reducing Ore

During the operating period not all of the feed material is separated into its two component isotopes. A considerable proportion of it goes astray and has to be recovered from the interior walls of the containing vessel and from parts of the operating equipment. The energy of the ionised particles is also so great that those which go astray and which impact on various parts of the apparatus combine either chemically or physically with the metals forming those parts. The result is that a large operating area must be provided for washing and cleaning operations, in which all parts of the operating apparatus are cleaned with steam, acid and in some cases, by an electro-stripping process.

The product is an acid solution of uranium compounds containing large quantities of iron, nickel, copper and a number of other metallic elements. All of these must be separated from the uranium before it can be recycled in the process, or before it can be passed on to the final concentration stage. Also, the uranium must be converted to the chemical compound which has been found most suitable for feed to the process. This compound must be of the highest degree of purity, completely dry and very finely pulverised.

Every Trace of Uranium Conserved

One of the interesting things about the chemical operations is the precaution which is taken to salvage every possible grain of uranium whose 235 content has been enriched. It is only by such precautions that the plant capacity is kept up to rating. Some of the precautions are interesting. One is the recovery of splashings from the uniforms of operators in the chemical area. In the cleaning process there is inevitably some splash, and drops of uranium solution are absorbed by the operators' clothing. To recover this, the uniforms are washed every day in small laundries in the individual buildings. The wash water is saved and treated for the recovery of its uranium content.

In the separation apparatus a large number of carbon parts are provided to protect the metal structures from beam impingement. These become impregnated with uranium atoms which impact on them. They are collected and sent to the incinerator building to be burned in an atmosphere of oxygen. The small amount of ash which results is treated for uranium recovery. Whenever a metal part is replaced, the old part is dissolved in acid which is then treated to recover any traces of uranium. The salvage operations are an extensive and important part of the plant operation. Even the ventilating air from the chemical areas is treated for any uranium-bearing entrained dust it might otherwise carry from the building into the outside atmosphere.

Future of Atomic Power

The future of atomic power must be considered from two points of view : (1) Its military aspects ; and (2) its industrial potentialities. The military situation may be summed up as follows :—

There is an abundant supply of uranium ore in almost every part of the world, and the supply of other raw materials for the production of bomb explosive is equally abundant. Scientific principles underlying the production of bomb material are known or are readily available to any nation desiring to make use of them. The secrets of the detonating mechanism are known to several nations : England and Canada and possibly to Denmark and France. These nations have also had an opportunity to learn the engineering secrets and production methods that have been tried and found successful in the United States. Therefore, the only thing which would prevent them from producing atomic bombs is their inability to construct and operate production plants.

Those other nations which have not been permitted to learn about this work are handicapped further by the need of working out the necessary designs. The magnitude of the construction work involved, and of the detailed design work which the construction of a new plant would require are such that it would probably take foreign nations from 5 to 10 years to produce an atomic bomb ; 5 years for those nations which have had access to the engineering secrets, and 10 years for those which have not. There is a further question whether any of these nations could in fact support an atomic-bomb programme.

As to the future of atomic power, all sorts of prophecies have been made that we are on the threshold of an atomic era in which unlimited quantities of energy will be available at low cost for industrial, commercial and domestic purposes. A piece of 235 the size of a pea, buried in front of one's doorstep, will heat the house during its entire life ; a slightly smaller piece will operate an automobile ; aeroplanes and steamships will be

similarly fuelled. These are some of the prophecies that have been made.

Power Installations Possible

It will probably be several generations before this millenium is reached, and probably no reader will ever heat a home or run an automobile by atomic power. However, it is believed that large power installations can be developed even within our lifetime. There will come a time, for instance, when the economies resulting from the continued plant operation and from the adaptation of new research discoveries will result in atomic power attaining a parity with power obtained from fuel. One pound of uranium 235 or plutonium will develop as much heat as 1,500 tons of coal. At 35s. per ton this establishes a cost of £2,625 per lb. for the 235, and this figure must be attained before the atom, as a power source, will be competitive with coal.

This goal is still a long way from being attained. When it is, it does not mean that all our steam-generating plants will be scrapped or become obsolete. Electricity is still the form in which energy will be distributed for domestic and industrial purposes. The steam turbine or gas turbine will still be the common type of prime mover. Boilers for the generation of steam will still be in common use, but they will be considerably different from the present ones. Furnaces will not be required, nor coal- and ash-handling equipment. Boilers will probably consist of a high-pressure drum with an immersed lump of atomic material evaporating water by the heat of the atomic fission. Of course, boiler feed pumps, steam piping, feedwater regulators, water-treating systems and most other auxiliaries will continue to be necessary. It can be shown that in the production of atomic energy about 94% of the invested capital of an electric utility system will continue to be useful; the other 6% will have to be replaced by atomic-energy apparatus. This is not a very gloomy picture as far as the electric investment outlook is concerned.

The liberation of energy incident to the production of plutonium has also been suggested as a means of cheap power. Such plants, or piles as they are called, do involve important health hazards and they will probably be constructed only in locations remote from inhabited areas and surrounded by large tracts of vacant land. A considerable amount of engineering work should be done to develop the energy liberation of this process. At the present time, the piles are water-cooled and although a great amount of heat is liberated in the process, the heat level is extremely low, being of the order of the heat which is rejected from steam-turbine condensers. It will be necessary to increase the pile temperature to a heat level, 700° F. or higher, sufficient to generate high-pressure steam or to heat gases hot enough for utilisation in a gas turbine.

These engineering developments are of considerable magnitude. However, they can be undertaken and solved within a reasonable period of time provided the problem is approached in an intelligent and co-operative manner so as to stimulate engineering initiative and the investment of capital in the atomic-energy industry.

Some Recent Appointments

Dr. A. G. Ramsay, B.Sc., Ph.D., A.R.I.C., has recently been appointed Manager of The Mond Nickel Company Ltd. Refinery, at Clydach, near Swansea. Dr. Ramsay assumed this appointment on May 1, and succeeds Mr. I. A. Bailey, who has become Managing Director of Henry Wiggin and Company, Ltd., a subsidiary of The Mond Nickel Company Ltd.

Prior to his present appointment with The Mond Nickel Company, Ltd., Dr. Ramsay was production manager, and also a member of the Management Committee of I.C.I. Metals, Ltd., Birmingham. For several years he was general manager of the Landore (Swansea) works of that company; thus he goes to Clydach with an intimate knowledge of local conditions in general, and, in particular, of neighbouring industries.

Dr. O. Sussman has resigned from the Board of Directors of The Anglo Metal Company Ltd., with effect at the 30th June, 1946. As from 1st July, 1946, Mr. H. K. Hochschild and Mr. C. Sussman are appointed Directors of the Company.

Dr. R. H. Barfield is joining the research and development staff of Wild-Barfield Electric Furnaces, Ltd. He will be relinquishing his position on the research staff of the National Physical Laboratory to take up this new appointment. Dr. Barfield has made many contributions to scientific developments during the past twenty years, particularly in relation to radio engineering, and in his new sphere his experience will be applied to some of the developments which have taken place in recent years in new methods of heat-treatment and other processes.

Mr. J. J. P. Mackenzie has been appointed by Hadfields, Ltd., as sales representative in the Yorkshire area. He takes over from Mr. Beaumont Sykes who is leaving the company shortly.

Dr. C. E. Homer has been appointed to the metallurgical staff of the Research Department of the Tube Investments Group of engineering industries. He is a graduate of the University of Birmingham and has previously been associated with the Tin Research Institute; Vickers-Armstrongs, Ltd., Castle Bromwich; and latterly with Darwins, Ltd., of Sheffield.

Plastic Coating to Protect Tools

A. C. Wickman, Ltd., Coventry, inform us that they will shortly deliver their "Wimet" tipped small tools with a protective plastic coating, covering the tungsten carbide for prevention of damage to the cutting edges during transport and subsequent storage by the user. The coating is easily removable without affecting the tool and can be re-used. Further details of this feature will be published in the near future.

Because of expanding business, and to give better service to local clients, David Brown and Sons (Huddersfield), Ltd., have opened an office at 71 and 72, Atlantic Chambers, 7, Brazennose Street, Manchester, 2; telephone, Blackfriars 0011.

Erratum

On page 23 of May issue the price of the book entitled "Electron Optics and the Electron Miscroscope" is given as 40/– net in Britain; this should read 60/– net in Britain.

Institute of British Foundrymen

THE 43rd annual conference of the above Institute, to be held at Birmingham from June 18 to 21, is arranged by the Birmingham, Coventry and West Midlands Branches. In addition to the usual technical sessions, which have been carried on during the war, social functions have been resumed, and, as in pre-war days, a programme is specially arranged for ladies accompanying members to the conference. A further return to pre-war facilities is the works visits, which complete a full and interesting programme; it is noteworthy that a banquet is a further reminder that the war is over although apparently the aftermath has no terrors.

Council and Committee meetings will be held on the afternoon of June 18 and in the evening members and their ladies are invited to a reception by the Birmingham Branch at the Botanical Gardens, Edgbaston. The annual meeting will be held at the Grand Hotel on the morning of June 19, at which awards will be presented and the new President, Mr. D. H. Wood, will deliver his address. After lunch Sir Claude D. Gibb, C.B.E., M.E., M.I.Mech.E., will deliver the Edward Williams Lecture; his subject being "Castings and Weldings." The remainder of the afternoon and the morning of June 20 will be devoted to technical sessions, at which the following papers will be presented for discussion:—

Session A:—"Control Tests for Grey Cast Iron," by Sub-Committee J.S.C. of the Technical Council. "Fluidity of Molten Metals," by K. L. Clark, Naval Research Laboratory, Washington, D.C. (American Foundrymen's Association Exchange Paper).

Session B:—"The Influence of Raw Material on the Properties of Whiteheart Malleable Cast Iron with Special Reference to the Influence of Residual Elements," by The Malleable (Metal Compositions) Sub-Committee of the Technical Advisory Panel to the Director for Iron Castings, Ministry of Supply. "The Annealing Rate in Gaseous Malleablising," by D. M. Dovey, M.A., and I. Jenkins, M.Sc. "Note on the Process of Gaseous Malleablising with Special Reference to the Use of Steam," by D. M. Dovey, M.A.

Session C:—"Standard Test Bars for the Non-Ferrous Foundry," by F. Hudson. "The Production of 'Grand Slam' Bomb Castings," by B. Gray, B.A. "Hot-Blast Cupola Design," by E. Longden, M.I.Mech.E. "Experiences with Balanced Air Feed in the Cupola," by T. H. Taft and H. A. Hallett. "Some Observations on Iron and Steel Castings Specially Resistant to Wear," by M. Ballay, D.Sc. (French Foundrymen's Technical Association Exchange Paper). "Heat Treatment of Grey Cast Iron for Relief of Internal Stress," by P. A. Russell, B.Sc.

Works Visits

Visits have been arranged to W. and T. Avery, Ltd.; Sandwell Casting Co., Ltd.; British Piston Ring Co., Ltd.; The Mond Nickel Co., Ltd.; Thos. Perry, Ltd.; Coneygre Foundry, Ltd.; Austin Motor Co., Ltd.; Humber, Ltd.; The British Cast Iron Research Association Laboratories; Morris Motors, Ltd.; Alfred Herbert, Ltd.; Imperial Foundry Co.; John Wright & Co., Ltd.; F. H. Lloyd & Co., Ltd.; and Birmingham Aluminium Casting (1903) Co., Ltd.

Edward C. Greig

THE announcement of the death of Mr. Edward C. Greig on May 13 would come as a shock to his many friends especially as his retirement from his position as Chief Labour Superintendent of the United Steel Companies, Ltd., had been announced just two months previously. Educated in Glasgow, the late Mr. James Henderson appointed him, in 1898, chief chemist of the Frodingham Iron & Steel Co. In 1907 he was made mill manager and subsequently steelworks manager of the Frodingham works.

About this time he took a great interest in labour problems, both locally and nationally, and in 1922 he went to the then head office of the United Steel Companies (Steel, Peech & Tozer) to become Labour Manager. In 1925 he became works manager of Messrs. Steel, Peech & Tozer and in 1929 he took up the position of Chief Labour Superintendent from which he retired in March at the age of 68.

Mr. Greig became a member of the Iron and Steel Institute in 1917, and in 1919 a member of the Iron and Steel Trades Employers' Association of which he was, for a long time, chairman of the Wages Committee. He was a representative of that Association on the Central Council of Wages Association, and was vice-president and chairman of No. 3 division of the Iron and Steel Trades Employers' Association.

To his colleagues his loss is more personal, they have lost not only a very good friend but a wise counsellor; his activities, however, were considerable and his loss will be felt in a much wider sphere, where his abilities and kindliness were well known.

TANGYES, LTD., of Smethwick, have re-opened a branch office in Manchester. It is located centrally at 5, Cross Street, and the district manager is Mr. J. G. Benwell. The telephone number is Blackfriars 5320.

SIR STANLEY V. GOODALL, K.C.B., has been appointed chairman of the new Council of the British Welding Research Association. The Council, which is the governing body of the Association, and over which Sir Stanley will preside, is composed of: Mr. J. H. Baillie, vice-chairman of British Constructional Steelwork Association; Prof. J. F. Baker, O.B.E., M.A., D.Sc., of the Chair of Mechanical Sciences, Cambridge University; Mr. Henry Berry, M.P., chairman of Metropolitan Water Board; Sir Leslie Boyce, K.B.E., chairman and managing director of Gloucester Railway Carriage and Wagon Co. Ltd.; Mr. Ivor R. Cox, managing director of Metropolitan-Vickers Electrical Co. Ltd.; Mr. Geo. R. Grange, a director of Alexander Stephen & Sons, Ltd.; Lieut.-Col. Sir John Greenly, K.C.M.G., C.B.E., chairman of Babcock & Wilcox, Ltd.; Mr. V. E. Greenwood, a director of Murex Welding Processes, Ltd.; Sir William J. Larke, K.B.E., director of the British Iron and Steel Federation; Dr. A. McCance, F.R.S., deputy chairman and joint managing director of Colvilles, Ltd.; Com. Sir Robert Micklem, C.B.E., R.N., deputy chairman and managing director of Vickers-Armstrongs, Ltd.; Dr. J. H. Paterson, managing director of Arc Manufacturing, Ltd.; and Mr. R. B. Shepheard, chief ship surveyor, Lloyds Register of Shipping.

MICROANALYSIS
CHEMICAL AND PHYSICAL METHODS
APPARATUS METALLURGICAL APPLICATIONS TECHNIQUE

A RECENT minimum estimate given to us of the cost of maintaining a student for a single session at one of the major universities was £275. That this is by no means over-generous is clear from the statements just made in Parliament regarding scholarships. Now let us see whither this leads us. To acquire a primary degree necessitates three years (at the least), while a research degree will require a further one or two years. The cost quoted above, too, represents only a session of 24 weeks. Presumably, in the intervals, a student must still live, and will therefore still remain a financial liability to someone. So that we can safely assume that the cost, either to the State or to an individual, of producing a qualified chemist may well be over £2,000. Now let us follow this qualified (but untrained) chemist when he applies for a post with the Corporation of X, where he will be required to supervise and control water or coal or gas production, or some of the other multifarious activities associated with public utilities. He is very often expected to have experience. But, passing that hurdle, he is very lucky if he is offered £350 a year. And, on his way home in the evening, he may, perhaps, stop and pass a welcome tip to his co-employee, the modern version of the crossing-sweeper, who, alas, is forced, simply because he has not had an expensive training, to exist on a wage—a pittance—of £7 a week.

Microquantitative Analysis of Sodium and Potassium

By F. Burton

(Continued from previous issue)

Even more than in the case of sodium, a wider range of reagents has been proposed for potassium than is generally recognised. This article discusses the reagents and methods which are of importance.

II. POTASSIUM

THE reagents most frequently employed in the determination of small quantities of potassium are probably sodium cobaltinitrite and chloroplatinic acid. An attempt is made here to outline some of the numerous methods which involve the use of these reagents, together with those which are less commonly used.

Sodium Cobaltinitrite

The dipotassium sodium cobaltinitrite precipitate has been fully investigated by Robinson and Hauschildt.[34] From their investigations it was confirmed that the composition of the precipitate can vary, depending upon the conditions of precipitation, the most important being the sodium and alcohol concentrations of the precipitating medium and the temperature at which the precipitation is carried out.

The usual method of preparing the sodium cobaltinitrite reagent is that of Adie and Wood in which the reagent is obtained by mixing a solution of sodium nitrite with a solution of cobaltous acetate in acetic acid. Precipitation by this reagent usually gives high results and it is obvious that a reagent containing less extraneous material is desirable. This has been accomplished by Wilcox,[35] who employs an aqueous solution of trisodium cobaltinitrite as the precipitating reagent, the reaction being carried out in a nitric acid medium. The precipitate obtained by this method is crystalline and heavy, is of apparently constant composition and is independent of the sodium concentration. Either a gravimetric or a volumetric finish using potassium permanganate in acid solution may be used.

One of the best known methods is that of Kramer and Tisdall[36] who decompose the precipitated dipotassium sodium cobaltinitrite with alkali, oxidise the liberated nitrite with potassium permanganate and back titrate the excess with sodium oxalate.

The disadvantages entailed in this method are outlined by Brown, Robinson and Browning[37] who have published an alternative procedure. The precipitated complex salt is dissolved in ceric sulphate and the excess is back titrated with ferrous ammonium sulphate using o-phenanthroline ferrous complex indicator. An average

34 Robinson, R. J., and Hauschildt, J. D., *Ind. Eng. Chem., An. Ed.* 1940, **12**, 676.

35 Wilcox, L. V., *ibid.*, 1937, **9**, 136.
36 Kramer, B., and Tisdall, F. F., *J. Biol. Chem.*, 1921, **46**, 339.
37 Brown, Robinson, and Browning, *Ind. Eng. Chem., An. Ed.*, 1938, **10**, 652.

error of 2% is reported in the determination of amounts of potassium over the range 0·2 to 1·0 mg.

A variation of the ceric sulphate procedure is presented by Kaye.[38] Potassium iodide is added to the excess ceric sulphate and the liberated iodine is titrated with standardised sodium thiosulphate using starch indicator. The obvious advantage of this method lies in the sharpness of the endpoint even with very dilute solutions enabling 0·02 to 0·12 mg. of potassium to be determined accurately.

Silver cobaltinitrite, the reagent introduced by Burgess and Kamm[39] has been used by Robinson and Putnam[40] for the determination of potassium in fresh water. A colorimetric determination of the nitrite content of the precipitate is made with sulfanilic acid and α-naphthylamine by which method as little as 0·002 mg. of potassium is determinable.

A rapid and accurate volumetric method using this same reagent is due to Klein and Jacobi.[41] The nitrite of the dipotassium silver cobaltinitrite precipitate is liberated with hot alkali and the precipitated cobalt and silver hydroxides are dissolved in sulphuric acid. Using ferroin as indicator, the solution is immediately titrated with ceric sulphate at room temperature. These authors prefer silver cobaltinitrite because of its greater insolubility, constant composition and easy manipulation of the precipitate. The error for amounts of potassium from 0·1 to 0·2 mg. does not exceed 2%.

Ismail and Harwood[42] also precipitate potassium as the double silver potassium cobaltinitrite and complete the determination either by estimation of the silver or the nitrite content of the complex. Silver is determined by the Volhard method and although the reported results are consistent, this method is only practicable with amounts of potassium greater than 0·1 mg. The volumetric determination of the nitrite content of the precipitate is essentially that of Bennett and Harwood[43] which employs ceric sulphate as the titrant and erioglaucine as indicator. The method affords satisfactory determination of amounts of potassium ranging from 0·03 to 2·0 mg. with an accuracy of 1%. Amounts down to 0·01 mg. can be determined but the accuracy is reported to be considerably less.

Colorimetric and Photometric Methods

A colorimetric method applicable to the microdetermination of potassium in biological materials and possibly to analysis of feeding stuffs and soils is described by Eden.[44] The method—a modification of that of Jacobs and Hoffman[45]—depends upon the emerald green complex developed by cobaltous salts with choline hydrochloride and sodium ferrocyanide. A wash solution of 70% ethyl alcohol is used in this method, although Kaye[38], who some years earlier used organic solvents both alone and mixed with water, found that they did not improve results. However, the method is reported to give constant and reproducible results within a range of 0·1 to 0·3 mg.

Sideris[46] determines potassium indirectly by precipitating with sodium cobaltinitrite and measuring in an optical colorimeter the amount of cobalt in the acid solution of the precipitate with the use of nitroso-R-salt (disodium salt of 1-nitroso-2-hydroxy-3 : 6 naphthalene-disulphonic acid) in buffered solution. Nickel and ferrous ions which interfere with the colorimetric determination of cobalt are eliminated in the procedure, which is suitable for the determination of potassium in biological materials. The same author has also adapted his method to the photoelectric colorimeter.[46a] Potassium within a range of concentrations 0·5 to 15 micrograms is successfully determined by measuring the intensity of the stable red dye of the nitroso-R-salt which is indirectly related to the amount of potassium in the precipitate.

Potassium dichromate is used in the photometric estimation of potassium described by Wander.[47] The precipitate of dipotassium sodium cobaltinitrite is used to reduce a definite volume of standard potassium dichromate solution resulting in the formation of yellow to green solutions. The absorption of these reduced solutions is determined with a photoelectric colorimeter and the amount of potassium in the sample obtained by reference to a standard calibration curve previously determined. An average error of about 2% is reported in the determination of 1 to 7 mg.

Other colorimetric procedures[48] which may be utilised after precipitation of the potassium with sodium cobaltinitrite include direct development of the cobalt chloride, development with dimethyl glyoxime and sodium sulphide, development with salicylic acid and α-naphthylamine and development with cysteine hydrochloride and hydrogen peroxide.

Chloroplatinic Acid

Cimerman, Wenger and Rzymowska[49] state that a gravimetric procedure using chloroplatinic acid is the most accurate method for the determination of 0·5 to 1·0 mg. of potassium.

Schohl and Bennett[50] have modified the classical chloroplatinate precipitation procedure and developed a volumetric method based on the following reactions :—

$$K_2PtCl_6 + 6KI = K_2PtI_6 + 6KCl$$

$$K_2PtI_6 + Na_2S_2O_3 = K_2PtI_4 + 2NaI + Na_2S_4O_6$$

The method is suitable for the determination of 0·4 mg. or more of potassium and is reported to have an accuracy of 2%.

A rapid volumetric method which gives results of high accuracy is presented by Bullock and Kirk.[51] Potassium is precipitated as the chloroplatinate, which precipitate is dissolved in water. Metallic magnesium in neutral solution is employed to reduce the platinum in the precipitate to the metallic state and the liberated chloride is titrated with silver nitrate using dichlorofluorescein indicator. Amounts of potassium within the range 0·5 to 1·0 mg. are determined with an error of 0·5%.

The titrimetric micromethod for the determination of sodium, potassium and chlorine in biological material due to Clark and co-workers[8] has already been mentioned. After removal of the chlorine in the sample by treatment with sulphuric acid, potassium is determined in the digest by precipitation as the chloroplatinate. Reduction is effected with the use of sodium formate and the liberated chlorine is determined manometrically.

38 Kaye, I. A., *ibid.*, 1940, **12**, 310.
39 Burgess, L. L., and Kamm, O., *J.A.C.S.*, 1912, **34**, 652.
40 Robinson and Putnam, *Ind. Eng. Chem., An. Ed.*, 1936, **8**, 211.
41 Klein, B., and Jacobi, M., *ibid.*, 1940., **12**, 687.
42 Ismail, A. M., and Harwood, H. F., *Analyst*, 1937, **62**, 443.
43 Bennett, H., and Harwood, H. F., *ibid.*, 1935, **60**, 677.
44 Eden, A., *ibid.*, 1943, **68**, 167.
45 Jacobs, H. R. D., and Hoffman, W. S., *J. Biol. Chem.*, 1931, **90**, 685.
46 Sideris, C. P., *Ind. Eng. Chem., An. Ed.*, 1937, **9**, 145.
46a Sideris, C. P., *ibid.*, 1942, **14**, 821.
47 Wander, I. W., *ibid.*, 471.
48 Snell, "*Colorimetric Methods of Analysis*," Chapman and Hall, London, 1936.
49 Cimerman, Wenger, and Rzymowska, *Mikrochemie*, 1936, **20**, 1.
50 Schohl and Bennett, *J. Biol. Chem.*, 1928, **78**, 643.
51 Bullock and Kirk, *Ind. Eng. Chem., An. Ed.*, 1935, **7**, 178.

Dipicrylamine

Magnesium dipicrylaminate is used by Kolthoff and Bendix[52] in a photometric determination. The precipitation is carried out at 0° C. and the precipitate dissolved in a few drops of acetone. The transmission percent of the diluted solution is determined and the amount of potassium obtained from a working curve.

A modified procedure due to Amdur[53] enables the precipitation to be carried out at room temperature. Using an aqueous solution of lithium dipicrylaminate saturated at 25° C. with potassium dipicrylaminate, amounts of potassium ranging from less than 0·15 mg. to 80·0 mg. may be determined with only a small error, and determination in the presence of sodium as in blood plasma is possible.

Dworzak and Ballczo[54] also recommend magnesium dipicrylaminate and report very satisfactory results in the gravimetric determination of 0·2 to 0·4 mg. of potassium when the error is of the order of 0·5%.

Phospho-12-Tungstic Acid

Van Slyke and Rieben[55] determine potassium in biological material with phospho-12-tungstic acid. The sample is ashed after digestion with sulphuric acid and a solution of the ash in hydrochloric acid is treated with the precipitating reagent. The precipitated salt is subsequently washed, dried and weighed.

An alternative titrimetric method is also described. The washed precipitate is dissolved in excess sodium hydroxide in the presence of thymol-blue phenolphthalein indicator. After boiling to remove carbon dioxide, the excess alkali is treated with excess sulphuric acid Carbon dioxide is again expelled and the solution back titrated with sodium hydroxide. Recoveries of potassium added to serum gave results within 3% of the theoretical amount.

Sodium-6-Chloro-5-Nitrotoluene-3-Sulphonate

Wiggins and Wood[56] in a comprehensive study of oilfield water analysis have investigated the above reagent as a quantitative reagent for potassium. One colorimetric and two volumetric methods were devised which proved satisfactory for the determination of potassium within a concentration range corresponding to 10·0 to 40·0 g. of potassium chloride per litre.

In the colorimetric method, the precipitated potassium salt is dissolved in hot distilled water, diazotised and treated with alkaline R-salt. The colour produced is measured by direct comparison against known amounts of potassium.

The volumetric methods depend upon the reduction of the nitro group either with stannous chloride or titanous chloride. The excess reducing agent is determined by titration with iodine solution and iron alum solution respectively. Reduction and titration in each case is carried out in an inert atmosphere. Accurate results are obtained with the titanous chloride method within the concentration range 25 to 35 g. of potassium chloride per litre.

52 Kolthoff and Bendix, *ibid.*, 1939, **11**, 94.
53 Amdur, E., *ibid.*, 1940, **12**, 731.
54 Dworzak, R., and Ballczo, H., *Mikrochemie*, 1939, **26** 322.
55 Van Slyke and Rieben, *J. Biol. Chem.*, 1944, **156**, 745.
56 Wiggins, W. B., and Wood, C. E., *J. Pet. Tech.*, 1935, **21**, No. 137, 200.

A Device for Delivering Small Quantities of Liquids

FOR delivering volumes of liquid of the order of a cubic millimetre or so, a calibrated platinum loop may be used. The device described below, constructed in a few minutes from glass rod, was evolved for the same purpose. Although it may, of course, be broken, it cannot suffer appreciable distortion, with consequent change in the volume of liquid delivered. Handled pencil-wise, the device is particularly convenient for delivering measured and reproducible quantities of liquid on to a microscope slide or similar surface.

The delivery of a platinum loop depends on the manner in which it is withdrawn from the liquid; thus, it is reported* that one particular loop delivered 0·75 cu. mm. when withdrawn in a vertical plane, but as much as 1·2 cu. mm. when withdrawn in a horizontal plane. If the present device is used in the manner recommended, a spherical surface is presented to the liquid to be dispensed. This is probably why variations in delivery of the magnitude mentioned above could not be produced.

Construction is quite simple. One end of a length of 3 mm. glass rod is drawn down to a thread approximately 0·7 mm. in diameter and 14 cm. in length. A 10-cm. length of the thread is cut off, and, by careful heating in the flame, spheres having a diameter of approximately 1·3 mm. are formed at both ends of it. The thread is then nicked in the middle and broken in two. The two halves are held parallel between the forefinger and thumb, so that the spheres are opposite to one another and about 0·5 mm. apart. The free ends of the threads are then fused together and joined to the drawn-down end of the rod, as shown in Fig. 1. The other end of the

Fig. 1.—Delivering Device.

rod is then rounded, and may be bent over to form a hook, thus allowing the device to be hung on the rim of a beaker. Before use, thorough cleaning with chromic acid, followed by several rinsings with distilled water, should be carried out. Moisture is then removed by a strip of filter paper.

To transfer a solution or reagent, the device is held approximately vertical, and lowered until the spheres just touch the surface of the liquid, then lifted immediately. The sample is delivered by touching the microscope slide, spot plate, etc., with both spheres simultaneously, holding the device pencil-wise and

* A. A. Benedetti-Pichler, "*Microtechnique of Inorganic Analysis.*" Wiley, New York, 1942, p. 42.

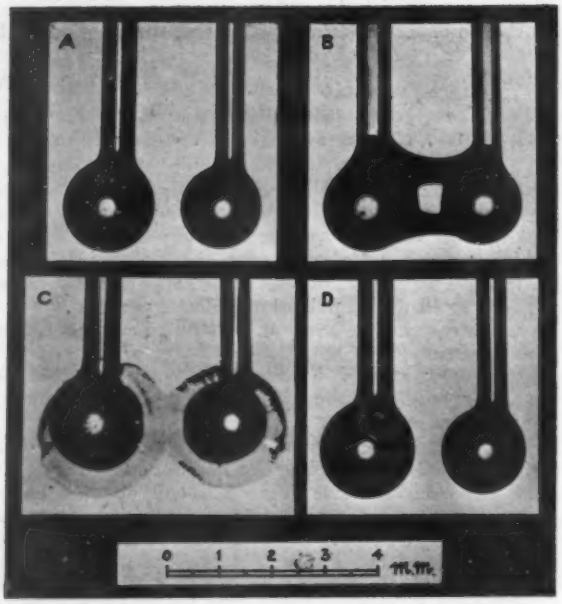

Fig. 2.—Stages in the Dispensing Operation. (a) Cleaned and dried ; (b) After touching surface of the liquid ; (c) Delivering on to a Microscope Slide ; (d) After removal from the slide.

withdrawing smartly within one second. It is important that the surface receiving the sample should be free from grease. Stages in the dispensing operation, photographed at low magnification, are shown in Fig. 2. Reproducibility is good, as is shown by the results in Table I. In obtaining these, ten successive deliveries of distilled water containing red dye were made on to a weighed microscope slide, which was then reweighed.

Fig. 3.—Device for Investigating Effect of Depth of Immersion.

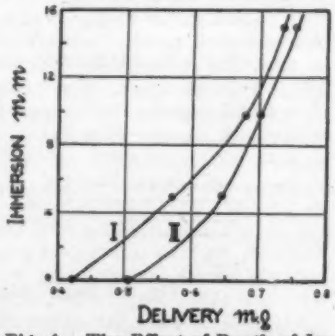

Fig. 4.—The Effect of Depth of Immersion upon the Average Delivery. Curve I, contact time, 1 second. Curve II, contact time, 5 seconds.

This was repeated six times, cleaning and drying the slide after each series of depositions.

TABLE I.—DELIVERY ON TO A MICROSCOPE SLIDE.

Series	Average Weight per Delivery, mg.	Difference from Mean, mg.
1	0·44	+0·01
2	0·42	−0·01
3	0·42	−0·01
4	0·43	0·00
5	0·44	+0·01
6	0·41	−0·02

To study the effect of variations in the depth of immersion in the liquid to be dispensed, the device was supported vertically in a cork, which was provided with three guide pieces made from glass rod, thus allowing the device to be presented squarely to the centre of the surface of the liquid in the vessel, as shown in Fig. 3. The head of the handle was carefully tapped until the spheres just touched the surface of the liquid ; this position was taken as "zero immersion." Any other desired immersion was obtained by depressing the handle by a measured amount.

Six series of determinations, each of 10 deliveries, were made successively at each of the selected depths of immersion, allowing a period of contact with the microscope slide of 1 sec. The results, which were reproducible to ± 0·03 mg., are shown in Fig. 4, Curve I. The effect of increasing the period of contact to 5 secs. is shown by Curve II. When the depth of immersion is more than a few millimetres, a period of contact of 5 secs. is visibly insufficient to permit complete drainage. Weighing experiments involving longer periods of contact gave results which were somewhat erratic, probably owing to evaporation losses. It is planned to follow the effect of further increasing the contact time by transferring standard solutions of dyestuffs to the slide, washing off the deposits, and determining the total amount of dye by means of a photo-electric absorptiometer.

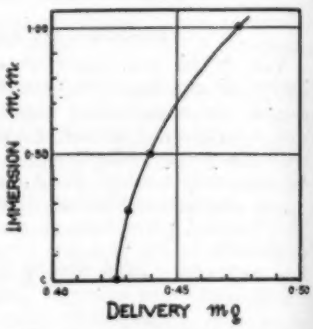

Fig. 5.—The Effect of Small Immersions upon the Average Delivery.

Depths of immersion insufficiently great to cause the spheres to be completely covered were next studied. Using a period of contact of 1 sec., the results shown in Fig. 5 were obtained. The degree of reproducibility was again ± 0·03 mg. It is interesting that increasing the immersion from zero to 0·5 mm. (approximately half-immersion of the spheres) caused an increment in the average delivery of less than 0·015 mg.

Later, it is hoped to examine other factors affecting the magnitude of the delivery, such as the diameter of the spheres and their distance apart.

Semi-Micro Quantitative Organic Analysis

IT is generally recognised that samples for normal analysis weigh about 500 mg., but applying micro methods of analysis involves samples weighing 3 to 6 mg. Considerable care is obviously necessary when adopting the latter method and, as will have been noted in this and previous issues of this journal, a special technique has been developed to reduce possibility of error to a minimum. To those not fully conversant with micro methods of analysis, semi-micro methods, which deal with samples of 20 to 50 mg., may serve as a stepping-stone to the more exacting micro-analysis methods.

By R. Belcher and A. L. Godbert, published by Longmans, Green & Co., Ltd., 43, Albert Drive, London, S.W. 19 ; price 10/6 net.

Soft Solder Selection Test

By E. E. Schumacher, G. M. Bouton and G. S. Phipps

WETTING data, along with information on melting point, solidification range and other physical properties are available on most of the tin-free and low tin alloy solders. While wetting data are of major importance, such data are difficult to apply to specific problems, and reliance must be placed on some direct test under carefully controlled conditions. A rapid test for such determination has been developed recently and is described.

The test consists in the vertical immersion of a pair of wires, 25 in. long, 0·040 in. dia., twisted so that adjacent turns are 0·5 in. apart to give the equivalent of a 1·0 in. pitch, at room temperature into the solder bath at a predetermined temperature, for a given time, and the measurement of the capillary rise of the liquid between the wires from the surface of the solder bath to the point of maximum rise. The specimens should be maintained in a vertical position after removal from the solder bath until solidification is complete. At the immersion end of the specimen, a minor amount of untwisting occurs during the test, which is relatively constant, but at a point ¾ in. from the surface of the test solder bath there is no detectable untwisting.

The rise of the solder, interpreted as a measure of its wetting ability under the conditions used, is due to capillarity and is controlled in the test by the temperature and nature of the solder, the diameter and thermal conductivity of the wires, the nature of the flux, the cleanliness of the wires, the time of immersion and the pitch of the twist. By maintaining other factors constant, the variation in wetting ability as a function of any one factor can be determined. Data obtained by this method are comparable, however, only to those determined in tests made in a like manner, and since the specimens are at room temperature before immersion in the solder, the test conditions tend to approximate to those encountered industrially when a soldering iron or other spot source of heat is used.

From *Materials and Methods*, 1945, Vol. 22, No. 5, pp. 1,407–1,410.

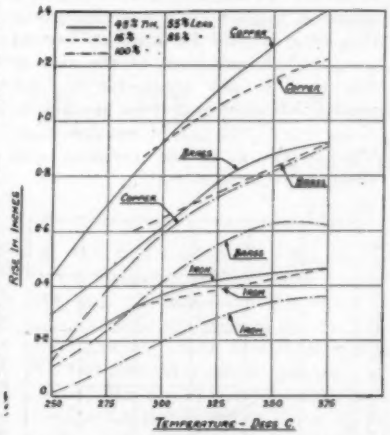

Fig. 1.—Extent of capillary rise in relation to temperature.

In Fig. 1 are given the results showing the extent of rise for 45% tin, 55% lead solder in relation to temperature using cleaned copper, brass and iron wires and zinc chloride-ammonium chloride flux. Results are also shown for 15% tin, 85% lead solder and for pure tin. The time of immersion in each case was 15 secs., and each value plotted represents an average obtained by the test of five specimens. From the curves it is evident that the temperature is an important factor in the wetting of the metals by solder, and in many instances the extent of rise is influenced to a greater degree by temperature variations than by changes in solder composition, and consequently in industrial soldering practice, a relatively high temperature should be used whose extensive wetting is desired.

In another test the extent of capillary rise between cleaned copper wires and between cleaned brass wires using zinc chloride - ammonium chloride and rosin fluxes, is plotted in Fig. 2 against the tin content in various lead-tin solders maintained at a bath temperature of 325° C. The alloys with tin content between 40% and 50%, those which have obtained wide use for general purposes, show approximately maximum rise. The solders rise higher than tin alone which is presumably the active wetting agent, probably due to the fact that the alloys have solidus temperatures that are lower than the melting point of tin. This test also shows the effect of the two general industrial types of fluxes.

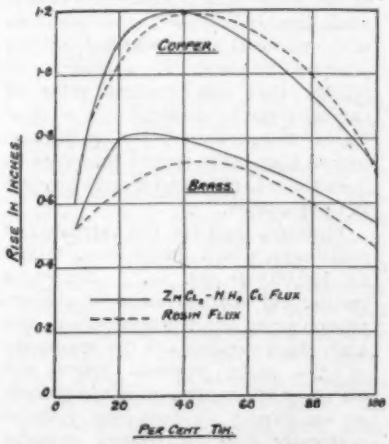

Fig. 2.—Effect of flux on capillary rise of lead-tin solder.

the corrosive chemical active zinc chloride-ammonium chloride flux and the non-corrosive relatively inactive rosin flux, on the extent of rise.

Although there was a relatively small difference in the extent of rise between the 15% tin, 85% lead and the 45% tin, 55% lead alloy, the difference in the quantity of solder that rose is rather large. Photomicrographs were taken of the circumferential spread at equivalent sections of test specimens made using these two solders at indicated distances from the level of the solder pool. A change in solder composition is noticeable between sections of the specimens taken at ¼ in. and

$\frac{1}{4}$ in. distances from the solder bath level, particularly in the 15% tin, 85% lead alloy. In the section taken at $\frac{1}{4}$ in. above the solder bath level, the composition represents approximately that of the bath itself, while at the $\frac{7}{8}$in. level the solder shown is a concentration of the lower melting point constituent in the alloy and approaches the eutectic composition. Since the quantity of low melting constituent is small, in a solder containing 15% tin, 85% lead, it follows that the amount of solder present at the higher levels of rise will be correspondingly small.

Factors Affecting the Rate of Extrusion of Aluminium Alloys

By T. L. Fritzlen

FACTORS, which are subject to technical control and which affect the rate of extrusion of aluminium alloys, are discussed in detail under three general headings: (1) Characteristics of the equipment; (2) size and structure of the ingot, and (3) temperature of ingot, cylinder and other parts. Well established facts in the extrusion of aluminium alloys are that the pressure and speed of extrusion are related and are experimental functions of one another, and that there is also a relationship between the temperature of the material being extruded and the pressure required to extrude. It follows that the greatest rates of extrusion can be obtained by the use of higher temperatures, but it is also known that other factors intervene to limit the rate of intrusion with increase in temperature.

Pressures used for the extrusion of aluminium alloys range from 50,000 to 150,000 lb./sq. in. Numerous mechanical characteristics of the extrusion press affect extrusion rate and such characteristics are the sensitivity of the main pressure valve and accuracy of pressure gauges, accuracy of equipment to heat the cylinder uniformly and to control cylinder temperature, and accuracy of means of measuring the speed of extrusion. As variables subject to control during the extrusion operation are pressure, temperature and speed, the characteristics of the press should be such as to enable close control of these variables. The strength of the auxiliary equipment such as tool holder, die, etc., may limit the maximum pressure that can be exerted during extrusion, and the length, contour and size of the extrusion die also affect the rate of production.

Another factor affecting the extrusion speed is the sharpness of corners or fillets in the extruded shape. Sharp corners limit the speed, otherwise cracks are produced. The size of the die opening also affects the rate of production. The area of the cylinder divided by the area of the orifice or orifices in the die gives the reduction of area from ingot to extruded shape which is termed extrusion ratio, and this ratio should be kept as low as possible. A reduction of the size of the ingot is not conducive to high production rate, and there are limitations also to the use of multiple dies. The structure of the material also limits the extrusion ratio.

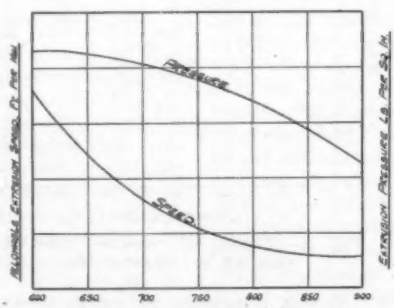

Fig. 1.—Approximate relation between temperature, pressure and speed in the extrusion of commercial aluminium alloys.

The size of the ingot is governed by the capacity of the press which in turn determines the maximum pressure that can be applied for any given cylinder diameter. A 6$\frac{1}{2}$ in. cylinder in a 2,500 ton press will allow the application of 150,000 lb./sq. in. on an aluminium alloy ingot, while an 8 in. cylinder in the same press will allow the application of 100,000 lb./sq. in. For the most efficient utilisation of a 2,500 ton press in the extrusion of strong alloys, the cylinder size and consequently diameter of the ingot should be selected to extrude at pressures as close to 150,000 lb./sq. in. as possible. For softer alloys, however, a 2,500 ton press can be used with larger cylinders, because the pressure required to extrude these alloys is lower. The length of the ingot to be extruded should also be considered. The longer the ingot, the surface in contact with the cylinder increases and correspondingly the friction increases.

Various aluminium alloys require different pressures to extrude, and the alloy composition and structure are important factors governing the rate of extrusion. Alloys in the as-cast condition tend to show a cored structure and this tendency, which is practically absent in alloys with low content of alloying elements is very pronounced in alloys with high copper or magnesium contents. The extrusion of as-cast ingots must be conducted at lower speeds than ingots having homogeneous structure. Lack of homogeneity may be minimised by a solution treatment. For high-strength aluminium alloys, a slow cooling after homogenisation permits lower extrusion pressures to be used.

The temperature of the material being extruded and the temperature of the equipment are very important factors governing the rate of extrusion. The speed of extrusion is dependent upon the temperature not only as a function of the pressure that will generate stresses, but also as to the amount of heat generated in the unit of time. In Fig. 1 is shown the effect of temperature in the range between 600° and 700° F. on the pressure and allowable speed in the extrusion of commercial aluminium alloys. For minimum pressure requirements, the material should be as hot as possible at the start of extrusion, and to obtain the maximum speeds of extrusion the temperature of the emerging shape should be as low as possible. Die, back-up and block temperatures are also important, especially when extruding shapes less than 0·064 in. thick.

Numerous other factors affect the rate of extrusion, such as condition of the die, mechanical troubles, tolerance allowed, and cleanliness of the metal, but the factors discussed have proved to be the most important ones and those requiring close control to obtain maximum production.

From *Metals Technology*, 1945, Vol. 12, No. 7, and A.I.M.M.E. Technical Publication No. 1851, pp. 1–8.

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Sleeves and Nozzles in Steel Pouring Practice

By M. Corbman

ONE of the primary functions of a nozzle is to soften at the pouring temperature of the metal in order to form its own seat when pressure is applied to the stopper rod assembly. If the nozzle is too refractory, as would be the case with sillimanite, this function would be lost. As most high alumina refractories do not soften until about 3000° F., they are not practical for nozzles.

High alumina refractories are difficult to fire to a dense body so that the cutting action of the steel on the nozzle would be as great or greater than with a conventional type clay nozzle. The range of apparent porosity of high alumina refractories is from 18 to 23%, whereas the porosity of clay nozzles ranges from 11 to 15%. The denser the nozzle, the more mechanical strength the nozzle will have, and will result in a less permeable surface of the steel to the nozzle, resulting in less cutting and erosion. However, where the nozzle is too dense, less than 11% porosity, its resistance to thermal shock is decreased so that spalling and cracks in the nozzle result.

The use of magnesite nozzles has been tried at Bethlehem Steel Works without success. The magnesite nozzle is mechanically weak and has poor spalling resistance. In the experiment the nozzle was preheated to a high temperature before starting to pour ingots, but the nozzle spalled badly in spite of the preheat. The inclusions which result from the spalled nozzle are more detrimental than the cutting and erosion from the clay nozzle.

The problem of sleeve refractories presents different factors than nozzles. The sleeve acts as an insulator for the stopper rod and the refractories should withstand the action of molten slag and metal for the duration of the pour. The sleeves should be more refractory than the nozzle. The sleeve refractory should not soften, but should remain rigid and dense so that cutting and erosion will be at a minimum. The sleeve refractory should be as dense as possible without spalling. Sleeves of apparent porosity between 11 and 15% have given better results than sleeves greater than 15% porosity. Below 11% porosity, they spall due to their high density.

Sleeves made from sillimanite or other high alumina super refractories should give good results where steel is being poured at either elevated temperatures or where the slags are exceedingly corrosive. These conditions are prevalent in electric furnace practice. The additional refractoriness of the high alumina super refractories will prevent the sleeves from burning through and severing the stopper rod. This retarding of the rate of erosion of the sleeve by the slag may be more costly, but this cost is offset by a better pouring practice.

A successful practice has resulted by using a sleeve, which is made by mixing the clay with a certain percent of carbon and burning in a reducing atmosphere. When these sleeves are used, only about half as much erosion takes place as the standard sleeve. The carbon present in the sleeve retards the rate of erosion so that the stopper rod is protected.

On the 7-ton basic electric furnace, in which is made special alloys and steels requiring high temperatures, the practice has been improved by using a clay graphite, sodium silicate mixture in all the joints of the rod assembly, and has reduced erosion of joints.

More refractory sleeves of the high alumina super refractories group are better than the conventional clay sleeves where corrosive slags and higher temperatures are prevalent.

Sleeves containing a small percent of carbon give less erosion than the conventional clay sleeves.

From a paper presented at Electric Furnace Steel Conference, Iron and Steel Div. A.I.M.E., Dec., 1945, at Cleveland.

The Effect of Various Elements on the Hot Workability of Steel

By H. K. Ihrig

A QUANTITATIVE, hot-workability test has been devised and used to determine the effect of various alloying elements on the hot-workability of steels containing them. The test method consists essentially of a rapid twisting of bars to failure in a furnace. A standard test piece, $\frac{9}{16}$ in. in diameter, prepared by forging and cold-drawing to size is inserted in a furnace with one end held in a chuck and the other end in a heavy dog. After soaking at the test temperature for 30 min. a torque is applied by the chuck and the dog makes contact with a stop which energises an electrical counter which records the number of revolutions made by the chuck. When the bar breaks in the furnace, the dog drops away from, and breaks contact with, the counter circuit, and an accurate count of the number of twists to failure of the bar under test is recorded. By running a series of tests on the same heat at different temperatures, curves can be drawn to show the temperature that gives the maximum number of turns and the type of curve for any particular heat. By correlation of the results obtained with mill operations, it is possible to predict whether or not a heat of steel will survive various hot-working operations such as forging, rolling and piercing, and what temperature will give the best results.

In killed open-hearth straight carbon steels containing from 0·04–1·04% carbon, under about 0·25% carbon the maximum ductility was found to rise with increasing carbon. Steels higher in carbon had their maximum ductilities at 1,290° C. or lower and as the carbon increased the ductility fell off rapidly with increase in temperature, Table I. In straight chromium steels, 12·47–22·38% chromium, the hot-workability decreased with high carbon contents, while nickel-chromium stainless steels low in carbon showed poor hot-workability.

In steels containing 0·005–12·88% manganese, manganese in general improved hot-workability, provided a uniform manganese sulphur balance was maintained. The 12·88% manganese, austenitic steel had much lower hot-workability than the ferritic manganese steels. Manganese above 0·41% appeared to give little additional improvement of hot-workability in chromium-nickel steels of the 18/8 type. In an 18% chromium, 8% manganese steel with no nickel, the hot-workability was poor.

A heat of steel with resulphurised ingots of sulphur contents ranging from 0·021–0·130% sulphur, with

From *Metals Technology*, 1945, Vol. 12, No. 7, and A.I.M.M.E. Technical Publication, 1932, pp. 1–29.

TABLE I.—Effect of Carbon on the Hot-workability of Plain Carbon Steels

Composition					Number of Turns at Different Testing Temperatures								
C	Mn	S	P	Si	1,095° C.	1,120° C.	1,150° C.	1,175° C.	1,200° C.	1,230° C.	1,255° C.	1,290° C.	1,315° C.
0·04	0·26	0·024	0·009	0·003	133	166	157	187	201	245	279	302	320
0·12	0·38	0·03.	0·01:	0·04	149	—	233	140	190	229	259	—	294
0·13	0·47	0·022	0·01.	0·03	217	243	270	288	295	312	332	344	357
0·20	0·48	0·03.	0·020	0·17	204	231	255	273	259	305	284	264	288
0·26	0·49	0·038	0·014	0·19	242	—	269	264	309	304	294	—	268
0·34	0·72	0·038	0·014	0·18	—	—	125	107	142	145	177	192	177
0·35	0·54	0·036	0·014	0·22	—	—	209	219	240	248	220	216	181
0·40	0·66	0·034	0·018	0·16	—	—	249	265	262	342	—	362	232
0·45	0·84	0·042	0·026	0·08	217	—	258	260	269	317	298	—	300
0·76	0·84	0·035	0·016	0·20	189	—	208	186	208	212	214	—	162
0·96	0·34	0·045	0·019	0·31	133	145	159	164	153	135	129	60	4
1·04	0·44	0·044	0·019	0·22	210	—	198	—	208	147	196	168	14

sulphur added, both as stick sulphur and sodium sulphite, showed the ingot lowest in sulphur to have the best hot-workability. In tests on steel containing varying sulphur, carbon and manganese contents, the steel containing 0·36% carbon, 1·74% manganese and 0·014% sulphur, gave the best results. Chromium reduced the red-short effect of sulphur and selenium and made ordinary stainless steels quite hot-workable, but selenium in 18/8 stainless steel lowered its hot-workability.

Low phosphorus steels, 0·013–0·064% phosphorus, showed normal workability curves, while silicon up to 0·2% had little effect in low carbon steel. Over 1·0% silicon, however decreased hot-workability in carbon steel and over 0·5% in austenitic stainless steels. Small amounts of nitrogen in carbon and low alloy steels did not affect workability, but in the higher chromium steels, ductility at high temperatures decreased with increasing nitrogen.

Tests on nine steels containing from 0·7–26% chromium showed hot-workability to decrease with increasing chromium up to 9% chromium, but above this amount the ductility increase, especially above 1,260° C. Pure nickel was found to be very ductile at high temperatures and nickel up to 5% in steels improved hot-workability at all temperatures. With 30·06 and 43·5% nickel, however, ductility was lowered probably due to the formation of austenite. Molybdenum reduced the hot-workability of carbon steel and molybdenum, silicon and vanadium had little effect on the ductility of 5% chromium steels. The hot-workability of the austenitic chromium-nickel steels at high temperature, however, was adversely affected by silicon, molybdenum, titanium and columbium.

A detrimental effect on the hot-working of low and medium carbon steels was obtained by the presence of 0·13% of lead. The comparative effect was not so pronounced, however, in steels containing approximately 1·5% of manganese and 0·12% of sulphur. Chromium had also a neutralising effect on the red-shortness tendency of lead. In steel containing as low as 0·057% of tin the hot-workability was lowered and the decrease obtained increased as the tin content was increased.

In general, most of the added elements studied tended to decrease the hot-workability of steels. Sulphur, silicon, molybdenum, tin and lead showed this effect in carbon steels, and nitrogen and chromium had a marked depressing effect on chromium steels. Manganese, nickel and above 9% of chromium improved the hot-workability of carbon steels. Manganese also neutralised some of the detrimental effect of sulphur. Oxygen, carbon, phosphorus, cobalt, vanadium and titanium individually had little, if any, effect. Room-temperature austenitic steels had much poorer hot-workability than room-temperature ferritic steels.

Ultra-Fine Surfaces on Metals

By K. Rose

RECENT years have brought a growing realisation of the value of ultra-smooth surface finishes for many uses involving moving parts. The automobile industry, bringing with it the petrol motor, has been one of the pioneers in the development of metal working techniques to produce smooth surfaces for engine parts, and without such processes the modern high-speed internal combustion engine would not have been possible. In addition to the automobile manufacturers, ball-bearing producers were also pioneers in smooth precision finishes. During the war the aircraft industry has profited from these developments, and in the post-war era many mechanical devices for domestic and industrial use will likewise be improved by the use of these better finishes.

Types of smooth-finish processes are grinding, lapping, honing, superfinishing and polishing. Although there is a considerable amount of disagreement about almost everything in connection with smooth finishes, particularly with regard to the nomenclature used for describing the different processes, several limitations are generally agreed upon as to the applicability of the various processes. All of the processes are intended for use after preliminary machining operations, and with all of them except grinding it is necessary that the finishing be done quickly, since the stone or other form of abrasive soon levels down the peaks in the surface roughness to provide sufficient bearing area, and no more work is done. Loose abrasives are unsuitable for work with soft metals, including some soft steels, since the work becomes charged with the abrasive. When soft metals are to be given a smooth finish, a bonded abrasive in the form of a wheel is used, and another process for finishing such soft metals as bronze, aluminium and silver, is diamond boring and turning.

Grinding is usually used as a preliminary to other surfacing techniques, and while ground surfaces can be produced by a skilled operator that will be as smooth as those by other methods, the latter do not generate the surface temperatures and pressures occurring with grinding, and so minimise the danger of damage to the work. When requirements call for dimensional accuracy within a few tenths, combined with geometrical accuracy, honing is most suitable. Honing can generate a surface, and so can correct taper, whereas lapping and polishing tend to follow contours. Lapping, manually or machine operated, produces work with the maximum degree of accuracy and surface finish, but metal removal is slow and the correction of inaccuracies is generally limited to high spots.

Superfinishing is a process in which metal removal is accomplished by abrasive stones under very light pressures and with multiple motions of work and abrasive and usually with the use of light oil as a lubricant.

From *Metals and Alloys*, 1945, Vol. 22, No. 1, pp. 70–75.

TABLE I.—FINE SURFACE PRODUCTION METHODS AND TYPICAL APPLICATIONS

Surface Roughness in Micro-inches	Production Method	Recommended Uses
500.....	Low grade machined surface.	For secondary parts not subjected to stress concentrations.
250.....	Medium machine finish or very rough ground.	For general use—surfaces under moderate tension, etc.
100.....	Smooth machine finish.	For main structural fittings, including highly stressed parts.
40.....	Very fine machine finish, or medium ground finish.	For highly stressed parts, except extreme tension members.
20.....	Fine ground, smooth reamed, coarse honed or lapped.	For rapidly rotating shaft bearings, extreme tension members.
10.....	Fine grinding or honing, lapping or buffing.	Interior honed surfaces of cylinders.
5.....	Honed, lapped, super-finished, very fine buffing or polishing.	For surfaces sliding across surface grain where lubrication is not dependable.

Dimensional accuracy can be obtained by using coarser stones with increased speed and pressure. Short, rapid strokes of the abrasive, with the light pressures used, prevent the formation of sufficient heat to cause surface damage. Polishing, which makes use of a loose abrasive and a soft lap such as cloth, felt or leather, is comparatively the cheapest form of surface improvement, but does not form any correction of surface inaccuracies. It is sometimes used to reduce wear on moving parts or to increase fatigue by removing machining marks. Other operations capable of producing fine surfaces are diamond turning and boring, carbide cutting, reaming, and burnish-broaching which is sometimes used for improving bore surfaces, especially with non-ferrous metals.

As important as surface finish is surface finish control. Since the dimensional variation called roughness is extremely small, its valuation is a difficult matter. In addition to the relatively fine-spaced surface irregularities known as roughness, there are those of greater spacing known as waviness. The unit for measurement of height of roughness is the micro-inch, 0·000001 in. Height may be specified as maximum peak to valley height, average peak to valley height, or average deviation from the mean surface. Roughness and waviness widths may be specified in inches and roughness measurements are usually made across the predominant surface pattern or in the direction which gives the maximum value. When only one number is used to specify height or width of irregularities, it indicates the maximum, and any lesser value is permissible. Maximum and minimum values may both be specified. Fine surface production methods and typical application are given in Table I.

Determinations of surface roughness may be made by comparison, either visual or by feel, by sound or by direct reading instruments. Visual methods making use of magnification are of use in the laboratory and for research work. A method developed recently is to use a replica of the surface made by coating the area with a film-forming material, stripping it off, mounting and projecting a large image on to a screen. Direct reading devices are the Profilometer and the Brush Surface Analyser. In the former instrument a diamond-pointed tracer moves a coil in a magnetic field to produce a small voltage, varying as the displacement of the tracer and after amplification the result is read on a meter as the root mean square value of roughness in micro-inches. With the Surface Analyser a very sensitive piezo-electric crystal converts the motion of the tracer into electrical energy, which in turn is continuously recorded on a chart by a stylus.

Cutting Tools for Various Metals

IN a contribution by the Management Division, and presented at the F. W. Taylor Memorial Session on Management, at a recent meeting of the American Society of Mechanical Engineers, Prof. O. W. Boston* naturally directed particular attention to Taylor's achievements, based on experiments carried on almost continuously over a period of 26 years, from 1880 to 1906. His work on the subject of cutting metals became a classic. Its clear and thorough presentation, with results so readily available for practice, won the confidence and admiration of readers. His work resulted in metals being machined faster and at less cost.

From 1898 to 1900, the Taylor-White process was crystallised. Previously only Mushet steel and carbon tool steel had been used. He considered the period up to 1894, the era of plain carbon tool steel. This type had been used for centuries. From 1894 to 1900, he considered the era of self-hardening steel, called Mushet steel, developed in 1868, by Robert Mushet, of Sheffield, England. From 1900 on was the era of high-speed steel. The faster and heavier cuts obtainable with this new steel revolutionised the art of cutting metals. New and more powerful machine tools were needed to use the steel to advantage and an immense impetus was given the improvement of equipment and method. Through the publication of Taylor's investigation, industry, in general, was stimulated to adopt his tools and methods.

Since Taylor's time, much further progress has been made in metals machined and in cutting tools. Many variations in high-speed steel have been presented to industry. Most of the high-speed-steel metal-cutting tools in use up to World War II were made from steel of the 18–4–1 type (18% tungsten, 4% chromium, and 1% vanadium). This type of high-speed steel, similar to "Taylor's best," has been the most generally used of all types, and excellent tools are consistently made by many manufacturers. There was a slight demand during the prewar years for the so-called low-tungsten high-vanadium type of high-speed steel, containing about 14% tungsten and 2% vanadium. This high vanadium was also added to the 18% tungsten steel and, in some instances, the vanadium was increased to 3 or 3½%. These tools gave increased hardness, toughness, and abrasion resistance required for machining very hard steels where slightly higher cost over the 18–4–1 type of steel was warranted by better performance.

In 1915, a cast non-ferrous alloy, superior to high-speed steel, was introduced. This type of cutting tool is still undergoing continuous improvement. It permits cutting speeds from 25 to 100% faster than the regular high-speed-steel tools on many classes of work. In 1928, a high-speed steel containing cobalt was announced as a super high-speed steel. This material, containing 4 to 12% cobalt, is capable of doing heavy-duty work, and some hitherto impossible work, such as cutting 13% manganese steel. In 1928, also, a new non-ferrous cutting-tool material known as sintered tungsten carbide was developed. It had a hardness much greater than any former

* "On the Art of Cutting Metals," *Mech. Eng.*, 68, pp. 304–306.

tool, except the diamond, and it had a high resistance to heat and abrasion.

In 1930, a sintered tantalum carbide for machining steel, otherwise quite similar to sintered tungsten carbide, was developed for the machining of steel. In more recent times, sintered tungsten carbide, sintered tungsten, titanium, and tantalum carbides combined, and sintered combined carbides of these metals have been developed in a number of different grades, each suited to a particular class of work. These metals are capable of operating at speeds 3 to 6 times those generally used with high-speed-steel tools.

Because of the shortage of tungsten, the development of the molybdenum high-speed steels was carried to a final conclusion, and these steels became generally used. In some, the tungsten was completely replaced by approximately 8% molybdenum, while in others the tungsten was only partially replaced. These steels are also made containing cobalt for heavy-duty work. Each of the several types of high-speed steel must be given its own individual type of heat-treatment in order to secure its best performance. The use of these steels became mandatory during the war and many people accustomed to using the 18–4–1 type had difficulty in developing a heat-treating practice suitable for the molybdenum types of steel.

The 18–4–1 type of high-speed steel or the molybdenum type is recommended for general use on all round work for the continuous-chip materials of the relatively low-carbon type. The high-carbon steels and tougher alloys call for the high-vanadium tool steels, such as the 18–4–2 types or the molybdenum-tungsten type. The nickel and nickel-chromium classes of steels seem to respond best to machining when cobalt-bearing tungsten and molybdenum high-speed steels are used. Stainless steels, manganese steels, and cast steels seem to be machined most economically with the high-cobalt type of tungsten or molybdenum steels. The high-vanadium type of tungsten steel has been found to give excellent results when machining the broken-chip type of materials, such as cast iron and nonferrous metals, where failure is usually by flank abrasion rather than cupping. Equipment for the heat-treatment and metallurgical control of tool quality has attained a high state of development and uniform tool performance is generally assured.

During the last few years a number of treatments of steels have been developed coming under the heading of cyaniding or nitriding, the so-called case treatments, as well as the plating of tools with chromium. These surface treatments usually must be given after each grinding if the lip face is removed. They are most effective in light operations or on abrasive materials.

Where Taylor had hard, medium, and soft steel, and hard, medium, and soft cast iron to machine, the present-day field covers literally thousands of types of steels and nonferrous metals, most of them in hot- and cold-finished bars and tubes, forgings, and castings. There are many carbon and alloy steels, plain and alloyed cast iron, malleable and pearlitic irons, and many high- and low-strength nonferrous metals of copper, aluminium, zinc, and magnesium, and a great variety of types and forms of nonmetallic plastics. Dozens of these metals are now being machined at hardnesses not even thought of by Taylor.

A whole new industry of cutting fluids for many specific purposes has been developed, whereas Taylor experimented only with dry cuts and with a tool flooded with water (soda water). Present cutting fluids consist essentially of soluble oils (to be mixed with water to form an emulsion), petroleum oils, compounded petroleum and fixed oils, and oils with additives such as sulphur, chlorine, graphite, etc.

Glycerine in Metal Treatment

By G. Leffingwell and M. A. Lesser

GLYCERINE has been found useful in metal treatment for quenching, plating, cleaning and polishing, as well as other applications. As quenching media for steels, glycerine-water solutions effectively bridge the gap between oil and water, and recent investigations on the heat-treatment of steel have shown that in mixtures of glycerine and water for quenching in the region between oil quenching and water quenching, the higher the ratio of glycerine to water, the milder the quenching. To produce high elasticity and maximum hardness, glycerine should be the major constituent, and elasticity is also enhanced by the addition of 0·25–4·0% of potassium sulphate, or still more by the addition of 1–2% of potassium chloride, or 1–3% of manganese chloride. With such media no dirt or deposits are formed on the metal, and traces of glycerine can be washed off easily.

There are a number of indications of glycerine's growing utility in the electrochemical treatment of metals, as illustrated by a recent patent specifying the use of this fluid in the electro-purification of lead, whereby lead, previously purified by the substantial elimination of copper, tin, antimony and arsenic, is formed into anodes, and an electric current is passed from these anodes through an electrolytic bath essentially of an aqueous-alkaline solution formed from lead oxide, caustic soda and glycerine, this last being present in sufficient proportion to increase the solubility of the lead compound in the electrolytic bath.

From *Materials and Methods*, 1945, Vol. 22, No. 5, pp. 1,432–34.

Glycerine has long been recognised as a valuable component of certain electrolytic solutions, especially those used in electro-plating processes, and recent applications of its use are in plating solutions for the electrodeposition of iron, chromium, zinc, nickel, manganese, molybdenum and tungsten. Investigations have shown glycerine to increase the throwing power of zinc sulphate baths within a wide range of current density, and moreover to improve the zinc deposits in quality; while in copper solutions at high current density glycerine reduced the throwing power, but increased it somewhat at low current densities. It has also been found that pure manganese can be deposited on iron, aluminium, copper, brass or bronze cathodes from manganese and ammonium sulphate solutions containing glycerine.

Glycerine likewise finds increasing use in metal cleaning following pickling and similar procedures. Added to rust-removing solutions, it also facilitates their use without affecting the derusting action. Mixed with soda crystals and sodium hydroxide in solution, it is also effective in cleaning alone.

Because practically all types of stainless steels are tough and do not lend themselves readily to mechanical polishing operations, the introduction of electrolytic methods was a marked advance in the treatment of metals. Many parts are so shaped that the usual mechanical polishing procedures are very difficult and sometimes impossible to apply satisfactorily, and electrolytic polishing, with its ability to get into corners, crevices and recesses of in-

tricate shapes is very suitable. The use of glycerine-phosphoric-acid mixtures for electrolytically polishing 18/8 stainless steel has been found to be satisfactory. It has been shown, using a photo-electric spectrometer, that a solution containing 42% phosphoric acid, 47% glycerine, 11% water, gives optimum polish. Electrolytic polishing methods utilising glycerine have also been developed for nickel, chromium, chromium steel and mild steel. Glycerine used in the anodic polishing of aluminium, magnesium and magnesium-base alloys gives substantial benefit.

Glycerine has an established place in standard methods for anodising aluminium by various processes. Used as an inhibitor in the several electrolytic solutions, it serves to prevent excessive etching of the metal, and helps to produce some white surfaces that are readily amenable to suitable colouring procedures. Another important use for glycerine in metal treatment is in soldering fluids and soldering pastes. The addition of glycerine to solutions for soldering stainless steel is recommended to give the flux sufficient surface tension to wet the surface thoroughly. Glycerine-containing solder pastes sometimes contain as high as 60% of glycerine. Glycerine is also finding increasing use in corrosion-inhibiting and anti-rust coating compositions.

Superheating of Magnesium Alloys

By N. Tiner

THE mechanical properties of magnesium-alloy castings are greatly improved by grain refinement and at present considerable attention is being paid to methods of obtaining fine-grained castings. One method of achieving this end is the use of proper melting and superheating techniques. In this investigation, the principal factors concerning the effect of superheating on the grain size of magnesium alloys were studied and attempts made to explain the experimental results in terms of a general hypothesis. Preliminary tests showed that when equal amounts of magnesium alloys were prepared from the same alloying materials, melted in graphite crucibles under the same thermal conditions, then cast into graphite moulds at a constant temperature, the resultant solids had the same grain size. It was therefore found possible, by maintaining all other factors constant, to vary thermal conditions of a melt and to examine the grain size of the resultant castings.

The alloys investigated, prepared with pure carbothermic magnesium and the necessary alloying elements, were melted under a flux in a graphite crucible heated uniformly by an electric resistance furnace. Samples were poured into cone shaped graphite moulds of two different sections, one large and one small. The pouring and the mould temperatures were about 700° and 300° C. respectively, as used in actual foundry practice. The presence of large amounts of β (Mg + Al) phase in cast alloys made the grain boundaries difficult to recognise and samples were therefore subjected to proper solution heat-treatment for a systematic determination of grain size. Proper sections of the homogenised samples were roughly prepared, polished and etched.

Grain-size testing was done by the transverse method. The sample was placed on a microscopic stage, the stage was moved for about 3 to 4 mm. successively in two directions perpendicular to one another, the grains traversing the centre of the cross hair of the eyepiece were counted, and the average number of grains per square millimeter were determined. Experiments were carried out to determine the effect on grain size of superheating temperature and time, of cooling from superheating to casting temperature, of structure of charge, of repeated superheating, of stirring, of superheating in vacuum, of bubbling various gases through superheated melts, of deoxidation of superheated melts before casting, of alloying constituents and superheating effect, and of superheating and undercooling.

In general, the effect of these various superheating conditions upon the grain size of magnesium casting alloys showed that the grain refinement by superheating was maximum at a certain temperature ranging between 850° and 900° C., and at higher temperatures of superheating there was a tendency for grain coarsening. Superheating molten alloys for several hours at low temperatures produced some grain refinement, but as the superheating temperature was raised to about 850° C., the time required to get the full effect dropped off to zero.

The superheating effect on grain size was gradually and completely neutralised by holding the molten alloys for increasing periods of time at low temperatures before casting. When the molten alloys were held at a given temperature for a sufficient length of time, the resultant grains exhibited, within certain limits, the same size whether the melts were brought to the desired temperature from lower or from higher temperatures. Stirring the molten alloys at about 800° C. caused a considerable grain refinement, but when stirring was carried out at 700° C. the resultant solid generally had coarser grains.

All other factors being practically constant, the grain size of an alloy varied within certain limits, with the structure of the charge before melting, and this variation in grain size was noticeable when the melts were superheated at low temperatures for short periods of time, but decreased and completely disappeared as the superheating temperatures were increased. Repeated superheating of the melts at elevated temperatures, 850° to 900° C. did not produce any appreciable change in the grain size of the resultant castings. It was also found that the effect of superheating on grain size could not be suppressed either by melting in vacuum or by treatment with a powerful deoxidising agent such as lithium or potassium. Bubbling nitrogen, chlorine and helium through

TABLE I.—AVERAGE GRAIN SIZE OF VARIOUS NORMAL AND SUPERHEATED MAGNESIUM RICH ALLOYS.

Alloy Composition Weight per cent.	Melt Temperature-° C.	Grains per sq. mm.	
		Small Section	Large Section
Pure Mg	700	Coarse Columnar Grains	
	850		
2 Mn	700	Columnar Grains	
3 Mn	700		
9 Al	700	35	25
9 Al	850	60	40
3 Zn	700	15	10
3 Zn	850	14	12
6 Zn	700	32	25
6 Zn	850	30	23
10 Al, 0·3 Mn	700	35	25
10 Al, 0·4 Mn	850	180	150
3 Zn, 1 Mn	700	15	12
3 Zn, 1 Mn	850	15	10
6 Zn, 1 Mn	700	30	20
6 Zn, 1 Mn	850	25	16
6 Al, 3 Zn	700	40	30
6 Al, 3 Zn	850	80	62
5·7 Al, 2·9 Zn, 0·33 Mn	700	35	30
5·7 Al, 2·9 Zn, 0·33 Mn	850	120	95
8·6 Al, 2 Zn, 0·3 Mn	700	40	32
8·6 Al, 2 Zn, 0·3 Mn	850	280	250

From *Metals Technology*, 1945, Vol. 12, No. 7, and A.I.M.M.E. Technical Publication, 1935, pp 1–19.

ELEKTRON MAGNESIUM ALLOYS

give Strength with Lightness

Comparative Finished Weights:

ELEKTRON	94 lb.
Aluminium Equivalent	155 lb.
Cast Iron Equivalent	388 lb.

Comparative Finished Weights:

ELEKTRON	33 lb.
Aluminium Equivalent	49½ lb.
Cast Iron Equivalent	130 lb.

● *Sole Producers and Proprietors of the Trade Mark "Elektron"* MAGNESIUM ELEKTRON LIMITED, Abbey House, London, N.W.1 ● *Licensed Manufacturers:* Castings & Forgings: STERLING METALS LIMITED, Northes Road, Foleshill, Coventry ● Castings: THE BIRMINGHAM ALUMINIUM CASTING (1903) COMPANY LIMITED, Bilton Works, Smethwick, Birmingham ● J. STONE & COMPANY LIMITED, Deptford, London, S.E.14 ● Sheet, Extrusions, Forgings & Tubes: JAMES BOOTH & CO. LIMITED, Argyle Street Works, Nechells, Birmingham, 7 ● Sheet, Extrusions, Etc.: BIRMETALS LIMITED, Woodgate, Quinton, Birmingham ● *Suppliers of Magnesium and Elektron Metal for the British Empire:* T. A. HUGHES & CO. LIMITED, Abbey House, Baker Street, London, N.W.1

superheated alloys had no great effect on grain size, whereas bubbling hydrogen produced a relatively fine-grain structure with numerous blow-holes. The superheating effect on grain size was observed in magnesium-aluminium or magnesium-aluminium-zinc alloys containing manganese or iron, but not in pure magnesium, magnesium-zinc, magnesium-manganese and extremely pure magnesium-aluminium alloys, Table I. The effect was found to be associated with a decrease in the undercooling capacity of the alloys. Attempts were made to explain the experimental results and a hypothesis that the superheating effect on grain size was related to the influence of solid soluble minor constituents upon the course of crystallisation was found to be satisfactory generally.

Fuel Efficiency in the Open Hearth

By B. M. Larsen and C. Siddall

IT is necessary to distinguish between the static elements, such as quantities of heat between various temperature levels, and the dynamic, such as rate of heat transfer and flame development. Comparison on a static basis will indicate the limiting efficiency of a fuel. With this, and considering also excess air leakage and regenerator design, analysis of fuel performance should be easier with respect to possible deficiency in fuel/air mixing rate or other dynamic elements. It provides, too, a basis for deciding the engineering possibilities of mixed gas—*e.g.*, blast furnace and coke-oven gas, and of furnace design.

As a first general proposition, it may be said that mixed producer and coke-oven gas, like oil or tar, gain appreciably from larger or deeper regenerators, together with relatively large gas checkers, and provision for leading a larger fraction of outgoing gases through the checkers. The best mixture in the case of coke-oven and blast-furnace gas is 35% of the former, and 65% latter, premixed and preheated. This has the highest capacity for returning waste heat to the melting chamber by regeneration, and therefore has maximum efficiency—superior to tar or oil. There are certain disadvantages in a mixed gas, but these are more than counter-balanced by advantages. It seems well worth while for American shops, where blast-furnace gas is available, to carry out practical tests in relation to furnace design and other factors. With the older type furnaces it may be better to preheat only the blast-furnace gas whilst the coke-oven gas or natural gas is injected at the ports. With a lower coke-oven gas content, say 10%, there may be appreciable increase in thermal efficiency and lower temperature at the end of checker zones. This mixture is also probably better for the earlier design furnace.

Some of the factors governing E_L (efficiency limit) are discussed. This, of course, is largely dependent on furnace design and use of the right gas mixture. Some of the earlier poor results were doubtless due to use of straight producer gas, for instance, and also to presence of large air leakage. For producer gas the ideal gas/air division of checkers should be about 25 : 75, instead of the more common 40 : 60 ratio. At present the gas chamber is not utilised effectively, whilst the air chamber may be overloaded. Most present-day furnaces have regenerators adequate for producer gas, and for oil or similar fuels. Although it is important to minimise air leakage, extra deep regenerators (preferably two-pass), and 18–20 ft. in checker work, may increase useful heat by 10%, and also accelerate melting rate.

It is emphasised that the present study cannot be of much practical help in solving many present-day problems of hearth design and fuel economy ; but it does suggest ways in which some experimental work could be undertaken in American shops on the lines of some of the best British or German practice, wherein some advance has been made in the economical use of producer gas from cheap coal, or coke-oven/blast-furnace gas mixtures. To assist in such practical study an analysis of the problem is presented on the lines suggested in a recent work published by the Am. Inst. Min. and Met. Eng., entitled, "Basic Open-Hearth Steel Making," New York, 1944 (Chap. XIV, pp. 404–445).

Compared with Great Britain, the U.S.A. has richer ore, and so produces relatively less fuel as blast-furnace gas, but recent developments urge a study of the most effective use of excess furnace gas. In regard to producer gas, reasons are given for using a minimum amount of steam in the producer, and for not overheating the gas in the open-hearth regenerator, together with the factors governing flame luminosity. The difference in optimum conditions of use of producer gas, and of the blast-furnace/coke-oven gas mixture are also noted. Straight producer gas, too, could be enriched by admixture of 5–10% coke-oven or natural gas.

Thermal efficiency of a fuel in the open hearth is reviewed under various assumptions, so that only two independent variables remain, namely : (*a*) type of fuel, and (*b*) depth of checker working. By means of an electrical analogue (K. Heindlhofer and B. M. Larsen, "An Electrical Analogue of the Flow of Heat in a Regenerator System," A.I.M. and M.Eng. Tech. Pub. No. 1798 Oct., 1944), the authors have determined quantities such as those given in subsequent tables for any combination of fuel and checker work depth, on the assumption that certain rate factors are constant :

(1) Percentage of heat input from fuel available to the hearth zone above the assumed critical temperature of 2,860° F. ;

(2) Preheat temperature of air or gas and temperature of outgoing products of combustion at the air inlets ;

(3) Approximate maximum temperature of brickwork in regenerative zones ;

(4) Volume of gaseous products per unit of high-temperature heat ;

(5) Efficiency limit (E_L) of the perfect regenerator.

On this basis, thermal calculations are made in respect to : (*a*) premixed and preheated blast-furnace/coke-oven gas, and (*b*) producer gas, producer gas plus 10% coke-oven gas premixed, natural and blast-furnace gas premixed, coke-oven/blast-furnace gas not premixed, oil and tar, coke-oven gas, and blast-furnace gas, as tabulated in following table.

Relative volumes of outgoing gases and peak temperatures of brick, together with respective areas of regenerator cross-section at each end, according to relative requirements for preheat between gas and air, are also tabulated.

In the subsequent discussion, A. J. Fisher, of the Bethlehem Steel Co., said that they used rather longer checker travel, 17 ft. each way in the two-pass. He stressed the importance of carbon radiation and of stabilising

From *Iron and Steel Eng.* Dec., 1945.

Depth of Checkers in Feet	Percentage High Temperature or Available Heat.		Mean Temperature of outgoing Gases at Air Inlets	
	20	12	20	12
Premixed c.o. and b.f. gas	70	57	770	1,140
" natural and b.f. gas	66	—	760	—
Oil	64	57	1,090	1,340
Producer gas+10% c.o. gas	62	56	1,120	1,320
C.O. (coke oven) gas	62	56	1,170	1,355
C.O. and b.f. not premixed*	59	53	1,130	1,320
Producer gas (straight)	55	50	1,360	1,525
B.F. (blast furnace) gas	53	—	960	—

* The c.o./b.f. not premixed is more suitable for older design hearths.

carbon content by additions in the form of pitch, lamp-black, or finely-ground coal, of 325-mesh or more (colloidal suspension). One of the future problems in open-hearth practice is to make a standard fuel with fixed carbon content. Another development is the supersonic oil atomiser in which the oil particles are broken up by high-frequency sound waves into their molecular constituents. Oil can thus be mixed with any gas. This opens up another new field for open-hearth fuels. The authors agreed that probably one of the main problems is flame luminosity, in which the injection of finely-divided oil or carbon is important.

Sealing Porous Castings

By P. S. Fuller

THE possibility of leakage in metal castings which must be pressure-tight has been substantially reduced during recent years by the use of plastic material which is excellent in heat and chemical resistant properties. During the war, increased demands were placed on foundries to produce pressure-tight castings of magnesium and aluminium alloys, copper, bronze, iron and other metals, and such castings were used in aeroplanes, ships and tanks, as well as in electrical and electronic devices. Micro-porous flaws, which are particularly prevalent in low density metals such as aluminium and magnesium, caused marked leakage of various liquids and gases in castings of those metals, and rejections of such castings reduced production considerably. Successful methods of impregnation for such castings were therefore of importance.

The impregnation of porous castings is not meant to be the medium of salving castings which are known to have foundry flaws, cracks and blow-holes. Its primary purpose is to ensure pressure or water-tightness by the elimination of microporosity in structurally sound pieces. In addition, it may be used after machining operations on castings which are not porous as delivered from the foundry, but which are opened up by the machining operations, as often happens where magnesium and aluminium are involved. The impregnation of all magnesium castings is generally accepted where specifications call for them to be pressure-tight.

The use of a chemical means of sealing castings is not at all new. Tung oil has been used for some considerable time and sodium silicate and phenolic thermosetting resins were both used in quantity until recently. The development of new alloys of magnesium and aluminium, their use in applications hitherto restricted to much heavier metals, the shortage of tung oil and dissatisfaction with other impregnants, have led to investigations for new sealing materials. After an intense study of such materials, Polyplastex, Inc., N.Y., found the right properties, both chemical and physical, in a modified, anhydrous, co-polymer, and this resin has been developed to suit any particular use. Its viscosity can be changed to satisfy a broad range of requirements and so ensure a proper flow into the pores of castings. The chemical resistance of the impregnated and cured casting is good and is completely unaffected by hydro-carbons, by other petroleum products, by glycols and by water or salt solvents. Acids and weak alkalies have little or no effect upon the cured impregnant. The cured resin in the pores of castings is tough, hard, yet resilient enough to allow for considerable expansion or contraction of the metal itself as various degrees of temperature up to 230° C. are encountered.

All castings should be completely machined before impregnation, as machining operations tend to open up internal pores. After machining, castings should be thoroughly cleaned by conventional solvent methods, heated to 120° C. to drive off the cleaning solvents and to eliminate any gases trapped in the pores, then cooled to room temperature. Three methods of impregnation are in general use, a pressure vacuum method, an internal pressure method and a vacuum method. After castings are impregnated by either method, they are well drained to remove sealing material, thoroughly cleaned and cured at a temperature of 120° C. with a time cycle dependent on the weight and thickness of the casting treated.

The need for impregnation is usually determined by testing a casting at a pressure 50% greater than the maximum operating pressure to be exerted on the part in service. Units can be tested by immersion in water or by coating with a soapy solution and applying air pressure to the inside of the casting. They can also be tested by applying water, glycol, mineral oil or mineral spirit under pressure to the inside of the casting and checking for leaks or seepage. It is usually desirable to test with the fluid to be used later on in the casting. By any of these methods, leaks are determined in the structure of the casting itself, are recorded, and then the weak points can be observed after the curing of the resin to determine definitely that the casting is pressure-tight. After impregnation, castings are inspected and retested in accordance with specifications. Generally, one impregnation is sufficient, but it is occasionally necessary to reimpregnate the casting in order to eliminate all pressure leaks.

From *Materials and Methods*, 1945, Vol. 22, No. 6, pp. 1,746–1,750.

Aluminium Beverage Cases

Aluminium is being gradually substituted for other materials used for beverage cases. Certain nesting features have been incorporated in the aluminium cases that allow for better piling, and also prevent movement when being transported. Their lightness makes them much easier to handle. The life of these cases is at least 15 years, while, in comparison, other cases are repaired at least twice in their 4½ years lifetime.

"The Aluminium Ingot" (Canada), 1946, No. 3, p. 7.